

# TRANSACTIONS

*of the*  
**American Society  
for Steel Treating**

---

## FEATURES

Dilatation of Steel During Quenching  
—G. M. Eaton 819

Selection of Case Hardened Steels for  
Highly Stressed Gears  
—H. W. McQuaid and O. W. McMullen 860

Study of the Burning and Overheating of  
Steel —W. E. Jominy 893

Steel Inspection Standards for Roller  
Bearings —A. S. Jameson 937

Constitution of Steel and Cast Iron—  
Section II—Part XIII F. T. Sisco 950

---



---

Vol. XVI No. 7  
December, 1929

---

# AGATHON ALLOY STEELS

*for speed with absolute safety*



**SPEED**—endurance—resistance to terrific stress and strain; they are today's demands.

And automotive manufacturers and designers, constantly seeking better, more durable materials to fill these needs—have found that Agathon Alloy Steels meet their requirements to perfection.

Agathon Alloy Steels are famous for their strength—dependability and longevity. For crankshafts—axles—pistons—rods—gears—steering knuckles—anywhere where strength, precision and endurance must be assured—these superb steels are the logical choice.

Our metallurgical laboratories are the finest in America. Here expert metallurgists are constantly striving to evolve even better materials.

They will gladly lend their assistance to your problems—at no cost or obligation on your part whatsoever.

**CENTRAL ALLOY STEEL  
CORPORATION**

*Massillon and Canton, Ohio*

**WORLD'S LARGEST AND MOST HIGHLY  
SPECIALIZED ALLOY STEEL PRODUCERS**

# AGATHON ALLOY STEELS



# TRANSACTIONS

## American Society for Steel Treating

VOL. XVI

DECEMBER, 1929

NO. 7

### DILATATION OF STEEL DURING QUENCHING

By G. M. EATON

#### Abstract

*Transient plastic flow during quenching is the mechanism which creates residual internal stress in steel. When the rolled surface of the steel is not machined off prior to quenching this transient plastic flow is severely localized in weak streaks in the surface material.*

*Martensitic dilatation aggravates this transient plastic flow, particularly in alloy steels.*

*The paper advances the theory that in heavy unmachined, quenched parts the passage through the austenite-martensite transition produces actual incipient failure which even if submicroscopic cannot be healed by subsequent tempering. This incipient failure does not reveal itself in the ordinary tension or torsion tests, but constitutes the cause of early failure in endurance tests.*

*Emphasis is laid on the need of industry for alloy steels in which the quality of the surface material of the rolled bar is improved till it bears the same ratio to core characteristics that exists in carbon steels.*

*Attention is called to the reasons why present endurance testing methods fail to evaluate steels which are in the condition noted and several methods are suggested for better evaluation.*

#### STEEL AS AN IMPERFECT MECHANICAL STRUCTURE

**W**HY does a given piece of steel fail in a specific service? There are three fundamental functions involved in the answer to this question.

##### 1. The ability of the steel to resist stress.

A paper presented before the Eleventh Annual Convention of the Society, Cleveland, September 9 to 13, 1929. The author, G. M. Eaton, member of the society, is associated with the Molybdenum Corporation of America, Pittsburgh. Manuscript received May 27, 1929.

2. The degree to which this ability to resist is uselessly burned up in opposing internal stress.
3. The stresses imposed by service conditions.

In discussing these three points in detail, particular emphasis must be laid on the fact that only one testing method has ever been devised which gives an absolutely reliable index of the ability of a given piece of steel to resist failure under the action of a specific stress cycle. This method consists of subjecting the particular piece of steel to the given cycle until it fails, or until it has run out a thoroughly satisfactory life without failure.

Since our discussion is directed toward situations where no satisfactory solution of the problem exists, failure is certain unless by our investigation we have advanced our art out of the specified class. Our test then is conclusive, but the patient is dead, and the piece of steel cannot be used for service.

By repeating the test many times on pieces of steel that are supposedly duplicates, we arrive at average results, which we assume represent a general characteristic of the steel, and we adopt a factor of safety (or ignorance) which, we hope, will enable us to eliminate failure or at least reduce it within limits which we may tolerate.

As soon as the first minute nucleus of failure has developed during production, on test or in service, the fate of the piece is sealed. It is therefore incumbent upon us to focus our attention on this first nucleus of failure.

If we study the fractured surfaces of the broken test pieces closely enough, we find that the location of the initial fracture is erratic, within limits which are dependent upon the conditions surrounding the test.

1. Why did the failure select this particular spot?
2. What could we have done in selecting the analysis, making the steel, forging the steel, treating the steel, modifying the design, modifying the service, etc., that would have delayed or eliminated the failure at this particular spot?

Answering the first of these two questions, the steel suffered initial failure precisely where it did occur, because at this location the relation of the ability of the steel to resist failure was most unfavorably related to the sum of the internal stresses and the imposed stresses. We emphasize strongly in this connection the possibility of

the actual starting of nuclei of fatigue failure in heavy alloy steel parts in the "as rolled" condition during heat treatment by the attack of transient thermal stress cycles on weak spots in the steel. This will be elaborated on in considerable detail later.

It is clear that the best available answer to the second question can be derived only through joint analysis by metallurgical and mechanical engineers, working with all the other thinking men engaged in all the phases of the production of the job.

This is brought out clearly when we realize that it is entirely possible for the initial failure to occur at a point where the total stress is lower than the maximum and where the resistance of the steel is above the minimum. If we assume that the conditions surrounding the service are beyond our control, and therefore cannot be modulated to ease the stresses imposed upon the steel by service, we can at once arrive at the conclusion that the steel failed because of local weakness. The failure was also vitally dependent upon the susceptibility of the weak spot to deterioration due to the action of the specific stress cycle involved.

#### TEMPERATURE-STRESS CYCLE DURING QUENCH

The discussion of weak spots in steel is utterly incomplete without some reference to heat treatment and to the transient stress cycle inevitably involved in the quenching and other heat cycles. Considerable literature exists on this subject. Most of this which has come to the author's attention, deals with the stresses which are residual in the steel at air temperature, after the heat treating cycles have been completed<sup>1</sup>. Some of it touches on the transient stress cycles set up by the rapid cooling action of the quench<sup>2</sup>. None of this literature, however, is satisfying to the author. He has tried repeatedly to write a word picture, from the qualitative standpoint only, of the transient cycle to which reference has been made, because of his firm conviction that the quenching process often inflicts minute permanent damage on the steel which cannot be healed by any subsequent tempering operations. This damage will usually have no noticeable effect in reducing direct tension or torsion values, but constitutes the actual start of fatigue failure.

<sup>1</sup>S. L. Hoyt: "Stresses in Quenched and Tempered Steel," *TRANSACTIONS, American Society for Steel Treating*, Vol. 11, April, 1927, p. 509.

<sup>2</sup>E. C. Bain and W. S. N. Waring: "Austenite Decomposition and Length Changes in Steel," *TRANSACTIONS, American Society for Steel Treating*, Vol. 15, January, 1929, p. 69.



We offer the following analysis with no defence of its detailed accuracy. It is only an honest trial to get the picture as clearly as the author has been able to, in an effort to judge as correctly as possible when quenching is the proper treatment and when it is wiser to resort to normalizing.

The following assumptions have been made to narrow the problem down to its simplest terms.

1. The sample piece is a solid cylinder in the "as rolled" condition, i. e., unmachined and so long that end effects need not be considered.
2. The problems involved in entering the piece into the quench are neglected. It is assumed that zero time is employed in getting the piece from "in quenching position but out of contact with the quench" to "immersed in the quench."
3. All consideration of convection currents is neglected. It is assumed that the heat flows radially away from the piece at a uniform rate in all directions.

The author offers for discussion the following picture of the cycle that will transpire under these conditions.

#### *A—No Surface Decarburization*

1. The surface layer cools with extreme rapidity<sup>3</sup>.
2. Skin tension builds up rapidly in outer cooling layer due to thermal coefficient of austenite. The degree of this tension is limited in a measure by the elastic limit of austenite, but it must be clearly realized that the speed of initial cooling is practically explosive and due to the reluctance of the material the stress conditions probably approach a dynamic set up.
3. Reactive compression builds up in the immediately underlying material, and this compression may or may not reach through to axis, dependent on speed of cycle and reluctance of material.
4. The skin tension is conducive to the austenite-martensite transition. The exact degree of tension necessary for this transition is not definitely known, and it is probably not a fixed degree, being probably tied up with the cooling rate and the resulting position of the critical point of transformation. We will assume however, that no martensite forms until the skin tension has exceeded the propor-

<sup>3</sup>H. J. French, G. S. Cook and T. E. Hamill: "Surface Cooling of Steels in Quenching." TRANSACTIONS, American Society for Steel Treating, Vol. 15, February, 1929; p. 217.

tional limit to a very material extent. (We realize that it is possible that tension below the elastic limit may provide the setting for the austenite-martensite transition.)

In other words, we assume that the austenite-martensite transformation occurs only in the presence of plastic flow, under the influence of tension.

5. Plastic stretch of the skin occurs.

6. This plastic stretch localizes in any and all weak nuclei or regions. This is an unfortunate condition, but we must face this fact as of great importance and as a highly probable agency for determining the precise location of initial fatigue failures.

7. As soon as the critical degree of skin tension is reached transition to martensite starts. We must note particularly that under our assumptions the localization of plastic flow in weak spots, also localizes the martensitic transition in some measure in the same weak spots. This will be referred to later.

8. Martensitic dilatation accompanies this transition.

9. This dilatation relaxes the skin tension.

10. This relaxation probably goes to zero and passes beyond into compression, to a degree involving plastic flow.

11. This compression builds up reactive tension on the material immediately inside the ring of martensite.

12. The depth of the zone subjected to tension is dependent on the speed of the cycle and the reluctance of the material to transmit forces.

13. The limiting speeds between which the actual speed must lie are:

(a) A speed so slow that the entire core inside the martensite ring is subjected to a more or less uniform negative pressure.

(b) A speed so fast that only a ring of differential thickness immediately within the martensite ring is actually subjected to tension.

(Items 14 to 23 inclusive are based on the assumption that the phenomena is in accordance with the first speed limit of item 13.)

14. An exceedingly high hoop compression would be imposed on the martensite ring in order to raise the tension on the entire core to the point necessary to permit the austenite-martensite transforma-

tion to invade the core. This condition appears highly improbable but we will follow the actions a little further.

15. The compression stress in the martensite ring would exceed the proportional limit of the material of this ring in spite of the fact that as this ring cools its proportional limit rises rapidly. This is due to the fact that the area of the martensitic ring is so small compared with the core area on which the martensitic ring is attempting to impose negative blind compression.

16. Plastic flow would set in as soon as the proportional limit is passed.

17. This flow would be radially outward as this is the only direction of freedom.

18. This flow again would localize in all weak nuclei and regions, and there would be a definite tendency for the martensite ring to separate from the core in any region of weakness in the binding material between ring and core.

19. While the outward plastic flow is in progress the temperature of the martensite ring is falling and the resulting contraction due to the thermal characteristic of martensite is combating the tendency of the martensitic dilatation.

20. This falling temperature of the martensite ring is also accompanied with a synchronous building up of the compressive proportional limit of the martensite ring.

21. When and if the negative pressure of the core reaches the intensity necessary (at the existing cooling rate) to induce the austenite-martensite transformation, the entire core undergoes this transformation.

22. Martensitic dilatation would occur in core.

23. This dilatation would relax skin compression through zero and into tension. If this tension exceeds the elastic limit, ultimate cooling will reduce it through zero to compression with reactive tension in core.

24. This skin tension would necessarily exceed the elastic limit with plastic flow to compensate for the plastic flow of the skin under compression noted in item 16. We are satisfied that the first speed limit of item 13 does not prevail.

25. At the second limiting speed of paragraph 13, which we believe is close to the actual occurrence, the speed of the martensitic dilatation in the outer ring is so great that tension or negative pressure is built up in the ring immediately underlying this outer ring



at a speed that is greater than the characteristics of the austenitic core enabling the material of the core to transmit the existing forces. In other words, we are dealing with strictly transient phenomena.

26. The tension in this underlying ring quickly reaches the degree conducive to the austenite-martensite transformation, and this underlying ring changes to martensite.

27. This degree of tension is probably not the same as that at which the outer ring went through transition, due to the radically different cooling rates of the two rings.

28. Martensitic dilatation takes place in the underlying ring.

29. This dilatation releases the compression in the outer ring, and since we have found plastic compression in the outer ring (item 10), the compression in the outer ring passes through zero and goes into tension.

30. At the same time the dilatation of the second ring imposes negative tension on a third ring.

31. The rate at which this tension is built up is still rapid because there are both the releasing compression of the outer ring and the expansion of the second ring to expedite the action.

32. The slower rate of cooling of the third ring, according to one school of metallurgical thought, depresses the temperature of the austenite-martensite transition, and at this lower temperature the speed of transmittal of force of which the material is capable, is greater.

33. As the diameter of successive rings decreases, the attack of the radially outward forces becomes more concentrated on the ring subjected to negative pressure.

34. As the diameter of successive rings decreases, the amount of material passing into tension on the outer rings, decreases the distance that each successive ring can expand as it passes through transition.

35. There may arise a set of conditions at some radial distance from axis which prohibits building up the tension which (at the cooling rate then existing) is sufficient for inducing transition to martensite.

36. Under this condition some portion of the ultimate core will not form martensite.

37. Departing now from the assumption that the material is an absolutely uniform structure and picturing it as it actually exists,

we must realize that there exist nuclei or regions that are less strong than the average.

38. The martensitic dilatation is then recognized as a dangerous enemy, since it forces the surface at any rate, and a more or less indeterminate amount of the underlying material to go through a compression tension cycle (with plastic flow at both extremes) and which is inevitably certain to localize in every individual nucleus or region of weakness.

39. For service involving either sustained load or load fluctuating within narrow limits, these damaged nuclei are of little importance; as they are usually of submicroscopic extent, and weaken the piece in a measure somewhere near their percentage relation to the entire section. Their importance is also some function of the manner in which service stress is distributed over the section.

40. The least damage is involved probably in a straight tension or compression service, where the stress distribution is practically equal over the entire section. Here, in a ductile material plastic redistribution of stress occurs, relieving the weak regions.

41. Under conditions of service subjected to a complete cycle of reversing forces, plastic flow may occur with every cycle in the effort of the material to redistribute stresses, and when this happens ultimate failure is inevitable.

42. We are then forced to the conclusion that in steels in which martensite forms readily in large quantity, a quenching cycle which would depress the transition point out of existence, permitting the transition to become austenite-troosite or austenite-sorbite would revolutionize the fatigue performance, since a great deal of the submicroscopic damage inflicted on the material would be eliminated.

43. This means that unmachined alloy steel could then be quenched in more massive sections than are now regarded safe for fatigue services.

44. In the discussion to this point, axial stresses have been neglected. This basis has been adopted for the sake of simplifying the analysis, which as we have seen is quite indeterminate at various points of the cycles even with this simplification.

45. There exist however, axial stresses of a high order.

46. In quenching a piece of infinite length (an utterly impractical assumption) there can occur no relative axial movement of any two particles of the material, if the material is absolutely uniform.

47. Since the material is not uniform, relative axial movement will occur at every nucleus or region of weakness.

48. This movement will go through a more or less complete cycle of response to tension, compression, tension, and maybe again compression.

49. If, as we have assumed, martensite can form only under some degree of tension which is a function of the existing cooling rate, the conditions for transition will be set up more certainly than would appear from the analysis which omitted axial stresses due to the fact that heavy axial stresses do exist.

50. Stepping over now into the field of practical quenching, the length of the piece becomes finite, and relative axial movement of various particles can and does occur.

51. We have seen no convincing statement of the length-area ratio necessary to build up the maximum axial stresses or plastic flow in a piece of steel which is quenched.

52. For whatever it is worth, we have been told by an engineer whose statements we have learned to respect, that the quenching of a 10-foot length of 90-pound rail would not be satisfactory or safe in an effort to learn the maximum internal tension in the cold rail after quench. He states that internal cold tension increases with the length quenched, up to the full 33-foot rail length.

53. With the limits between which the ratio of length to cross sectional area varies in practical quenching (omitting quenched wire) it seems safe to state that the stress conditions are not identical in any two contiguous differential axial or radial locations in the piece.

#### *B—Surface Decarburized*

When the surface of the quenched bar is decarburized, the thermal stress in the outer surface layer of the material is radically different from that occurring when decarburization is absent, as follows:

1. The outer surface cools, shrinks, and takes on tension with plastic flow, about as before.

2. Due to the fact that carbon is absent or at most is present in small quantity on the surface, little or no martensite can form at the surface\*.

\*If the statement is true it helps to explain the early failures attendant on surface decarburization. In attempting to verify this statement from the literature, we found we had stumbled into a controversial subject.



3. The ring of metal just under the decarburized material is in hoop compression due to the shrinkage of the outer decarburized ring, but practically no plastic flow can occur in this second ring, due to this hoop compression, as there is no direction in which the metal can flow; therefore all the plastic flow occurs as stretch of the decarburized ring. (The only plastic flow that can occur in the underlying ring is that involved in a possible local collapse of weak spots.)

4. The second ring meanwhile is cooling, shrinks, loses its compression, takes on tension with resulting plastic flow, and enters the austenite-martensite transition, with consequent dilatation.

5. Just at this point we meet the vital stress difference due to decarburized surface.

This dilatation occurs in the ring underlying the decarburized skin at a time when the decarburized ring is already shrinking and flowing plastically under tension due to its own temperature and thermal characteristic. The decarburized ring is therefore shrinking down onto an expanding core, and is subjected to a degree of plastic flow that is more severe than anything encountered in a similar quench of a piece which is not decarburized.

#### CHANGE OF LENGTH DURING QUENCH

While we do not maintain that the preceding analysis is closely correct even from the qualitative standpoint, we believe that the tendencies to which attention has been called are closely enough representative of what actually takes place to justify the statement that the length-time curves in the paper by Bain and Waring are not representative solely of precise fundamental characteristics of either the steels or the quenches involved, but they are representative also of the diameter and length of the pieces employed.

The overall changes of length measured to produce these curves are the average of changes which vary throughout the length and radius of the piece.

M. A. Grossmann suggested in his discussion that it would be desirable to conduct further experiments with specimens of different diameter. We concur with Mr. Grossmann, but consider it equally important to repeat some of the runs with specimens of different length.

## NEW TEST FOR LENGTH CHANGES DURING QUENCH

It seems to be logical to believe that the purpose of these tests can be achieved better by using a radically different test specimen and quenching rig. As we understand it, the fundamental reason for conducting these tests is to determine the cycle of change of length which a given steel will undergo under the influence of a known rapid temperature cycle. Certain prime fundamental facts are obvious for the ideal set-up.

1. The analysis of the steel must be known accurately.
2. The grain size of the steel must be known accurately.
3. The specimen must harbor the irreducible minimum of internal stress at the instant of quenching.
4. The temperature gradient throughout the body of the specimen must be at the irreducible minimum throughout the cycle.
5. The specimen and the entire set-up must be adapted to transference from furnace to quench with a very small drop of specimen temperature.
6. The set-up must record—time cycle, temperature cycle, length change cycle.

From these prime fundamentals we can deduce certain secondary or resultant fundamentals.

- a. From 3 and 4 above, the thickness of the specimen in the direction of major heat-flow must be as small as practicable.
- b. From 3 and 4 above, a circular shape is unavoidably selected.
- c. From 4 above, the heat flow must be limited to the greatest possible degree to the radial direction and this flow must be as uniform as possible in all angular directions in the radial plane.
- d. There is an apparent clash of prime fundamentals as between items 4 and 5 above.

This can be overcome, however, if we can provide additional thermal capacity of thermal flywheel immediately adjacent to the specimen.

- e. This thermal flywheel must be in intimate and uniform contact with the specimen; capable of giving up its heat readily and uniformly to the specimen; incapable of imposing sensible mechanical stress upon the specimen, and must retain these characteristics throughout the temperature range which is under investigation.
- f. The initial contact of the quenchant with the specimen must

be as nearly as possible uniform and synchronous over the entire exposed surface of the specimen.

g. The specimen must be kept free from oxidation.

Fig. 1 shows a section through the specimen and quenching device which we propose for the approximation of the fundamentals which have been listed. This device is described in considerable detail in the appendix. Briefly, the specimen is a thin-walled tube. This tube is mounted in an electric extensometer, and after being thus mounted it is heated for quenching in a molten salt bath. Thermal flywheel is provided by filling the inside of the specimen with molten salt which enters during the heating for quenching.

The quenching device consists of a cylindrical tank containing a measured amount of the quenchant. This tank is rotated by an electric motor to a speed sufficiently high to form a vortex in the quenchant in the axial region, and with practically vertical sides, and of a diameter sufficiently great to permit entering the specimen to quenching position with no contact with the quenchant. After the specimen is in position for quenching, the rotation of the tank is suddenly stopped by a powerful electric brake, and curved webs inside the tank throw the quenchant into contact with the specimen. It is entirely possible with this device to pass from "all out of the quench" to "all in the quench" in about a hundredth of a second if such speed is desired. The main function involved in speed of quench contact faster than this degree is the amount of money to be spent on the device.

The entire record of this instrument may be produced synchronously on the oscillograph, i.e., temperature, time, and change of length. With this set-up it will be possible to remove the specimen from the heating bath, enter it into the quenching device, and secure contact of the quenchant over the surface of the specimen in less than two seconds.

The records of temperature et cetera can start at the instant of starting to lift the specimen from the heating bath and continue without interruption till the specimen has passed through the entire critical range, and down to the temperature at which the salt inside the specimen solidifies. Below this range there is a possibility, with a very thin-walled specimen, that stresses imposed by the salt would introduce errors in the record of length change, unless the salt is decanted before reaching its freezing temperature. There may be



other heating baths that would overcome this characteristic of the molten salt bath.

It is theoretically possible that the specimen be heated inductively while in position in the spinning quenching tank instead of in a molten salt bath. This suggestion was made by J. W. Weitzenkorn, president of the Molybdenum Corporation. This method if feasible has some material advantages over the salt bath. This would make great changes in the apparatus, and this is discussed in some detail in the appendix.

Analyzing the cycle through which the specimen will pass, we find that:

1. The outer skin cools, shrinks, builds up hoop tension, undergoes plastic flow, and enters the austenite-martensite transition.
2. Martensitic dilatation occurs, releasing the skin tension, building up hoop compression in the skin and subjecting the inner skin to tension stretch and plastic flow, with resultant entry into the austenite-martensite transition, etc.

We are forced to confess that we are not quite as sure about all the plastic flow phenomena as in the case of a solid bar, but as far as we can see, the occurrences take place as noted. This is discussed in a little more detail later.

We wish to call attention to the unusual opportunity offered by this set-up for the investigation of the effect of mass with a given steel, by quenching specimens of successively increasing thickness and studying the three types of recorded data. By plotting the data on the basis of wall thickness, the curves can at first thought be extrapolated toward zero wall thickness, thus securing a line of information as to what takes place during quenching, about which at present we now know little or nothing. Unfortunately there is strong probability of something akin to a discontinuity or at least a radical change of characteristic as the thickness approaches zero so that this extrapolation would have to be handled carefully. It seems to us fair to say that with the proposed test we can learn more accurately than we now know the temperature at which the austenite-martensite transition occurs under the influence of various cooling rates.

We expressed some doubt as to the plastic flow occurring during the cooling of this specimen in the quench. There must be some thickness of wall so small that when the outer skin shrinks and takes on hoop tension, the thickness of underlying metal which undergoes

reactive compression is so small that it is mechanically unable to impose plastic flow on the outer skin. In this case the first plastic flow which occurs will take place on the inner portion of the wall and under the influence of compression. Later on in the cooling cycle these tension and compression stresses will of course be reversed, leaving hoop compression in the outer layers and hoop tension in the inner layers when the specimen is cold. Probably there will be no plastic flow in response to these final stresses. If this critically small thickness can be produced practically in the specimen, we have a good method for proving the truth or falsehood of the assumption that martensite forms only under the influence of tension plastic flow. For if this is true, no martensite will be formed in this critical thin-walled specimen. If martensite formation is not limited in this manner the entire analysis of transient thermal stress cycles falls to the ground.

This type of test also offers an attractive means for proving whether or not transient martensite will form during quenching of a steel which is very low in carbon, in the case of steel where mill decarburization exists at the surface. It is quite essential to realize that the results of a test of this type do not reveal an absolute characteristic of the steel which is entirely divorced from the shape and size of the specimen. The plastic flow is inevitably tied up with the thickness of the wall, and after this thickness is reduced on successive tests to a certain degree, the initial plastic flow of the outer skin begins to fall below the amount which occurs in a solid specimen of the same outside diameter. If our assumption of dynamic speed of the initial cooling cycle is correct, then the critical thickness referred to in the preceding sentence is probably quite small, though we cannot properly hazard a guess as to where this critical thickness lies. In fact we think it quite likely that it is different for different analyses with the same grain size, and also different for the same analysis with different grain sizes.

In case an investigation were being undertaken with this type of test, where the thinnest possible walls were to be explored, it might be advisable to change the shape of the curved webs in the quenching tank so that the impingement of the quench would depart further from the radially inward direction and toward a tangential direction as it is possible, with the webs as shown, to impose an undesirably high hoop compression in the specimen due to the initial impact of the

quenchant, and this might entirely invalidate the results on a very thin-walled specimen.

When investigating very thin-walled specimens lack of uniformity in the material of the specimen might invalidate the results. The value of the results of this type of test depend entirely on careful interpretation. Probably the major value would be, as is so often the case, in the comparison of the performance of various steels and grain sizes. We invite criticism on this proposed type of test.

#### THEORY FOR EARLY FATIGUE OF HEAVY UNMACHINED QUENCHED STEEL ALLOY PARTS

The main exciting cause for following this line of study is the hope of finding the reason for the early fatigue failure of quenched and tempered alloy steels in certain drastic services, particularly where large sections are involved and in which the rolled surface of the steel has not been removed. By going one step farther in our analysis we can reach a tentative theory.

In order to simplify the discussion of the mechanical phenomena attendant on cooling we have considered the actions as progressing in circular rings. This, however, is not in strict keeping with our assumption that martensite forms under tension and as some function of plastic flow. We have called attention to the fact that plastic flow localizes in all spots that are mechanically weak, and that therefore martensite also localizes in these same weak spots. This means heavier dilatation effects and in general an increased plastic disturbance of the material surrounding the weak spots.

It is probably useless to pyramid the analysis further along this line. But it seems essential to carry it to this point in order to bring out for discussion the following theory to explain why heavy helical springs and heavy leaf springs of alloy steel fail during fatigue tests under the effect of less cycles than carbon steel springs.

In the alloy steels in question, martensite forms more easily, more massively, and more stably than in the carbon steels.

In many and perhaps all of the alloy steels adapted to this service surface conditions are inferior to the carbon steels. (The conditions of surface to which we refer can be shown surely only by properly selected etching, as the visual appearance of the bar is not a safe criterion. Therefore, the alloy steel has had the candle burned at both ends, if we may be permitted the expression, since its inferior

surface has been subjected to more serious local disturbances, and these disturbances have actually started submicroscopic nuclei of fatigue fracture, which cannot be healed by any tempering operation.

We thus arrive at a conclusion which we believe is much more reliable than the full detail of the substantiating analysis. Until the surface of alloy steel is very materially improved, it is not conservative practice to subject large sizes, in the "as rolled" or unmachined condition, to drastic heat treatments and to put the product into services which involve such heavy fluctuating stress cycles as must be met by the severest helical and leaf spring applications.

As surface improvements are effected in alloy steels the sizes which can be handled conservatively, will increase. In smaller sections, notably automotive parts, the superiority of alloy steels, in the condition under discussion, has been proven beyond argument.

It is only fair to state that the practices which will improve alloy steel surfaces will doubtless also improve the surfaces of carbon steels. It remains to be seen whether the degree of improvement that can be made in the alloy steel surface will render its surface resistance to failure, which is available for service, greater than the corresponding value which can be attained in the carbon steel. Until this end is attained, carbon steel will maintain its supremacy for the class of heat treatment and service we are discussing.

#### THE IDEAL QUENCH

While our attention has been focussed on steel in the "as rolled" condition, we must realize that all steel that is subjected to drastic quenching undergoes transient plastic flow cycles which are more or less a function of martensitic dilatation among other exciting causes. The cleaner the steel, and the freer the steel is from localized internal stress, the less the danger of incipient failure occurring during the quench.

The practice of subjecting steel to drastic quenching and subsequent tempering is as old as the heat treating art. It is usual to hear and read of relieving stresses, residual after quenching, by the tempering operation.

We know that in many applications, the difference between an immediate temper and a temper delayed a few hours after quenching, spells success in the one case and failure in the other. The temper may relieve residual stresses, but no temper, soon or late, high or



low, ever relieved an incipient nucleus of fatigue failure, once that nucleus was fairly started even to a submicroscopic degree.

To us the inference is clear that the ideal quench is one which would produce directly the desired structure and physical characteristics and possibly even obviating the necessity of a tempering operation. We have no material technical objections to the tempering operation as such, but we do aspire to see localized transient plastic flow brought down to the irreducible minimum because localized transient plastic flow is the mechanism which builds transient and residual internal stress of the most dangerous character, because inflicted upon the metal least able to offer successful resistance. And finally the way to reduce this flow to a minimum is to adopt a treatment whereby the structure may pass through a transition from austenite directly to the finally desired structure and with no taint of transient martensite.

We are told that the molten salt bath quenching operation which is creating so much interest at this time, actually can be made to eliminate the martensitic structure entirely, the structure passing from austenite to the desired structure directly. We understand that the critical feature is the rate of cooling which is set up together with its effect in depressing the critical point. We sincerely hope that this is true, as it appears to us to offer a logical hope of opening doors for the application of alloy steels in cases where those doors are now closed. We refer to services such as heavy helical springs, heavy leaf springs, etc.

We have cast about considerably for a convincing method of proving whether or not the molten salt bath does actually escape the transient formation of martensite in fairly heavy sections, as any deductions based solely on an investigation of thin specimens does not satisfy us. Our proposed method for measuring length changes during the quench does not look very attractive for a molten salt bath, though it seems technically possible to adapt it to this type of quenchant.

The best arrangement we have been able to work out capitalizes on the well known deflection cycle followed by a railroad rail while cooling on the hot bed. The current explanation for one complete deflection cycle or wave of deflection is based on the passage through the martensitic structure, as follows:

1. The rail runs out on the hot bed with the camber that ex-

perience has shown brings it to the cold condition in the closest proximity to straightness.

2. The base being of less mass than the head cools more rapidly than the head and therefore contracts in accordance with its thermal characteristic.

3. This contraction produces camber, concave on the base side, in the rail.

4. As soon as the base cools to the critical degree the base enters the austenite-martensitic transition, with resultant expansion due to martensitic dilatation.

5. This dilatation produces a reversed camber, i. e., concave on the head side.

6. Then the head enters the austenite-martensite transition and the camber is again reversed, the base having passed from martensite to troostite-sorbite.

7. As the head later passes to troostite-sorbite, the camber gradually disappears and the rail when at atmospheric temperature is nearly straight.

Then if we quench a rail-shaped section in the salt bath which is reputed to avoid martensite, and it does this, we will find that we have eliminated the martensite deflection wave entirely. This is a very simple test to conduct, and would be quite convincing to the author, if the deflection cycle resulting could be explained entirely on the basis of the usual thermal characteristic of the metal in the unbalanced rail section.

#### AIR HARDENING VERSUS QUENCHING

We lean strongly toward air hardening processes instead of quenching operations, wherever it is at all possible to approximate the desired physical properties by air hardening, always remembering that our interest must center in the resistance to failure over and above that portion of the total resistance which is uselessly employed in opposing internal stress and which is therefore not available for service.

The outstanding safety feature of the air hardening process lies in the comparatively flat temperature gradient that is set up. As the temperature gradient flattens down, the amount of plastic flow is reduced, and after all is said and done, plastic flow is the enemy.

we must fear in connection with the transient conditions surrounding the heat stress cycles met in quenching steel. The most dangerous feature encountered in connection with plastic flow lies in the fact that plastic flow automatically and inevitably ferrets out every weak spot and localizes in these spots which are least able to resist the plastic flow and which by the action of plastic flow suffer a decrease in resistance to the stresses which are built up after plastic flow has ceased.

In this connection we repeat that in much of the plastic flow, the trouble may not come to a head while the flow is taking place, but the plastic flow plants the seeds which at a later stage in the cycle bear fruit in incipient failure.

We have seen the tendency of the thermal cycles set up during quenching, to locate distress leading to later failure in the weak regions of steel. Some of these tendencies doubtless exist in normalized parts though in far less degree, while the purpose of furnace annealing is, among other aims, to reduce thermal gradient and resultant plastic flow to a minimum.

#### INGOT SURFACE

Our consideration of this subject would not be sufficiently complete if we did not get as clear a picture as we can of the reason why streaks are present in the surface of rolled stock to a greater degree than they are in those parts of the stock just inside the surface.

At this point we wish to avoid any possible misunderstanding by defining what we mean when we refer to streaks. In this paper when we use the term "streak" we mean that structure or condition in steel which, on deep etching, develops an elongated longitudinal opening that resembles a crack in appearances. There may be absolutely no visual evidence of this on the surface before the etch, or there may be a longitudinal crack, or there may be only a stress concentration, or an elongated segregation or inclusion. We class here anything falling within this whole range of defect as a streak, because under the conditions which we are discussing, they all lead to much the same class of failure, differing only in degree of seriousness.

When steel is poured into the ingot mold, the metal adjacent to the walls cools in advance of that more remote from the walls. This metal, which cools first, promptly becomes viscous enough to entrap

nonmetallic inclusions to a much greater degree than occurs in the metal toward the core of the ingot. Also the first cooled metal is subjected to severe tension stresses. The fluted ingot was developed to reduce these stresses. We have often wondered just what the effect would be if the ingot mold were heated as hot as the safety of the mold as a mechanical structure would permit, before the steel is poured into the mold. From the standpoint of the life of the mold, this should be helpful, if the maximum temperature remains below the range in which scaling would occur, as it would reduce the initial temperature gradient through the wall of the mold and thus decrease the tension stresses imposed upon the metal of the inside of the mold during the cooling half of the temperature cycle to which it is exposed. We judge it would make only a small difference in the maximum temperature inflicted upon the inner surface of the mold and therefore only a correspondingly small difference in the ability of the metal on the inner surface to resist stresses.

It would delay the trapping of nonmetallic inclusions in the surface metal of the ingot, which is an end to be greatly desired in steel which is to be heat treated and put into service without machining to remove ingot surface metal.

It would at first thought seem to delay the final freezing of the ingot and therefore encourage more dendritic growth and segregation, which are of course hostile.

But where the failures of the finished product practically always initiate at the surface, we are perfectly justified in knowingly deteriorating the core if by so doing we increase the resistance of the surface to failure. This question of the relative degree of ingotism and segregation with a hot and a cold ingot mold however, is not as simple as it might appear. With a cold mold, the metal enters and fills the mold. The mold starts to heat up and it expands. The molten metal follows this expansion until it has built up a frozen wall strong enough to contain the fluid core. Then the ingot begins to shrink away from the mold and the resulting space between the ingot and the mold constitutes heat insulation which radically delays further cooling and thus encourages dendritic growth and segregation.

If the ingot mold is preheated to a temperature above that which exists in a cold-poured mold at the time that space begins to develop between the mold and the ingot, it seems logical to conclude that



throughout the cooling range the air space between the mold and the ingot will be narrower than is the case with the cold-poured mold, and this will tend to expedite cooling, with concurrent decrease in ingotism and segregation. We reiterate that the safety of the mold must not be neglected in this study.

These detail comments on the relative merits of a cold versus a hot ingot mold constitute a groping for possible improvement in ingot surface, with a glance at attendant gain or loss from other standpoints, and the subject is offered for discussion, without firm conviction on the part of the author who is not particularly skilled in this subject.

The importance of improving ingot surface metal is clearly shown when we realize that the surface of a rolled bar contains practically the entire metal of the longitudinal surface of the ingot. If any further proof of this demand is required for steel that is to go into service without machining and after drastic heat treatment, we find it in the growing practice of putting helical spring stock through the centerless grinder for the sole purpose of removing the inferior surface material. This inferiority of course involves decarburization as well as ingot inferiority. This is a procedure which we heartily advocate with the surface conditions now existing in helical spring stock for drastic service, but we hope it is only a temporary expedient and we believe firmly that a determined attack on the conditions causing the inferiority due to both ingot surface metal and decarburization would result in sufficient improvement of this critical metal to obviate the necessity of centerless grinding in many applications where it is now the only way to achieve successful service in the finished article. If the improved stock is more expensive of course the customer must pay, but the cost of centerless grinding gives a good margin for developing an improved surface in the "as rolled" condition.

We wish to avoid having anyone think that we regard this problem of making radical improvement in the surface metal of ingots as a new thought or that we consider the solution as a simple matter. Let us define the situation as clearly as we may.

A tremendous amount of attention has been paid to the quality of the ingot surface metal, and great improvement has been made, during the history of the steel industry. This development however, has not kept pace with other advances in the quality of steel, and we

refer here particularly to the general alloy steel development. In fact the surface of alloy steel as rolled is often inferior to the surface of plain carbon steel.

The designing mechanical engineer, in response to the demands of industry, calls for continually higher physical characteristics in his steels, and at the same time is forced by competition to narrow his factors of safety. He is enabled to do the latter with apparent propriety, because of his increasing knowledge of stress distributions, as he learns how to use the principles of elasticity, etc.

In a tremendous number of applications of alloy steel in industry, the maximum stresses are imposed upon the outer surfaces of the critical structural members and the metal in these surfaces has not been improved anywhere nearly as much as the average improvement throughout the entire body of the steel. It is therefore logical that we seek for improvement in the critical region of the new materials of construction.

#### FATIGUE TESTING

We must go further than the study of the steel making in our effort to improve the service results we are securing with modern steels. Let us consider some of the methods employed in investigating the properties of these steels, remembering that our discussion is largely centered on the ability of the steel to resist varying stress cycles and realizing further that we are also focussing on steel which is heat treated and put into service without machining.

Practically all of the fatigue testing with which we are familiar where a rotating beam or cantilever is employed, produces results which cannot be safely used as a guide for the designer, under the conditions which we are discussing. We do not decry this work where it is applicable, but we dare not use the results here.

There are two major reasons why this is an indisputable fact.

1. The test pieces usually employed are so small that in the act of machining them the greater portion of the residual stress, existing in materially larger sections, is automatically relieved. In this connection we are much gratified to learn that fatigue testing with samples some  $2\frac{1}{2}$  or 3 inches in diameter is now under way. This, however, is only a step toward much larger pieces, to which we must inevitably go before we can secure a thoroughly reliable basis for the interpretation and commercial application of the results of fatigue testing.

2. By the act of machining the test piece, the inferior ingot surface as well as the decarburized material is removed. We thus secure results which tell us the endurance limit of material which, in the class of application we are analyzing, is practically never subjected to the maximum stress.

It seems to us that great service would be rendered to industry by prosecuting fatigue tests along the following lines.

1. Prepare the test pieces as at present and then heat treat them in various cycles, and test them with no further machining or polishing. If the present form of test piece is too susceptible to distortion to permit this procedure, then adopt a sturdier form of test piece which will permit this important line of testing.

2. Develop drop forged test pieces and specially rolled test pieces, with their critical surfaces composed of ingot surface metal, and test these by present fatigue testing methods after they have been through various thermal cycles. These pieces to have no surface finishing operations.

3. Develop cast test pieces which are unmachined in the critical sections and test them, as cast, as annealed, etc., through the whole range of thermal treatment.

4. Develop torsional fatigue testing facilities for testing straight rolled stock, unmachined, under a cycle of varying degrees of torsion. In aiming toward the determination of the adaptability of the steel for helical spring application, an effort should be made to devise a setup whereby the stress concentrations fundamentally present in the helical spring should be approximated. If this last line of testing were faithfully carried out, any discrepancy between the results on test and the results in service would at once direct attention more forcibly than at present toward the practices of the spring maker. This is not done effectively at present, because the practices of making and testing steel for this important, difficult and unsatisfactory service are sufficiently open to criticism that we have the situation of the steel maker and the spring maker each holding the other responsible for the present intolerable conditions, while the testing engineer sits back and blames both. As a matter of fact, we are all party to the situation.

#### RESEARCH REQUIRED

If we may be so rash as to venture the opinion, we see great

possibilities of gain for some ambitious steel maker to conduct a comprehensive research of the way to improve the surface of stock that is to be heat treated and go into service without machining as well as of ways to prove the advance that has been made. Such a research followed with grim persistence is sure to evolve patentable procedure, which will pay the cost of the work many times over. It is not an easy problem and cannot be solved in a brief time; in fact it means a running study, consolidating one gain and immediately starting out on the trail of the next step, for industry will absorb each gain as soon as it has been made and will calmly ask for more. But the steel maker who will do this will lead in his art.

We are quite convinced that this research cannot advisedly be split up among industries, but to develop the maximum possibilities must be conducted under one guiding head who will view all phases of the development dispassionately, and be willing to find room for improvement wherever it may exist. Perhaps the logical place for this research is to be found at the Bureau of Mines.

This is a difficult and discouraging line of effort and we will be intensely interested in sitting back and watching to see whether any one has the courage to rise to the bait of future financial advantage.

#### BURIED WEAK SPOTS

Thusfar we have dealt with local weaknesses of steel most of which are traceable to the conditions existing in the surface of the ingot. There are, however, other vital classes of failure, directly traceable to local weak spots in the steel, which demand attention. These are due, at least in part, to the known inferiority of the metal at or near the core of the ingot, due to segregation and piping, etc. These are recognized as germane to our subject when we remember that the plastic flow incident to martensitic dilatation is liable to attack weak spots regardless of their location in the finished article.

It would be out of place in this paper to discuss in any great detail the heterogeneity existing within the body of steel ingots. But it is quite germane to note that internal rupture is present to a very disquieting degree. (See the three "Reports on the Heterogeneity of Steel Ingots"—British Iron and Steel Institute). We refer to this fact in order to bring out the thought that the shatters which we are wont to ascribe to the thermal stresses attendant upon quenching are probably often the reopening of ingot ruptures which have been



closed and perhaps partially welded during the working of the steel.

We are convinced that one of the outstanding needs of the steel industry is a greatly increased attention to the analysis of the phenomena of freezing as applied to ingots. This study involves very close mechanical analysis along with the metallurgical phases of the problem. In our judgment there are very great possibilities of developing steels with materially less volumetric change during the act of freezing and with reduced thermal shrinkage characteristic below the freezing temperature. This is another fertile field for research, where tremendous reward is attendant upon a successful solution.

We may classify failures due to buried weak spots as follows:

1. Failures which originate remote from the metal which was at the surface of the ingot and also remote from the surface of the finished article. The outstanding illustration of this class of failure is transverse fissure of railroad rails.

2. Failures which originate remote from the metal which was at the surface of the ingot and yet on the surface of the finished article. The oil well drill stem, other than which perhaps nothing in the steel industry today is more thoroughly an unsolved problem, illustrates this class of failure.

3. Then there is a class, lying in the border land between the two preceding classes, which from a tonnage standpoint is of comparatively small interest, but where the service is so vital, and the practice so hard on the steel, that it demands mention. We refer to the H-section connecting rod as used on many large modern steam locomotives. This does not quite fit our specification of no machining, but is germane to a study of weak spots in steel, as attacked by plastic flow.

We know of no other single failure of steel for which so many theories have been advanced, as those purporting to explain the development of transverse fissure, and we lack the boldness to expatiate on the theory to which we subscribe. There are, however, certain points on which there is fair agreement, as follows:

1. Transverse fissure originates in a weak spot in the steel.
2. Transverse fissure is a tension failure.
3. The structure of the rail passes through martensite as it cools on the hot bed.

If this were not so we could not account for the deflection cycle through which the rail passes as it cools.

4. When the rail is cold, the parts of the cross section most susceptible to transverse fissure are subjected to internal tension as a result of the temperature stress cycle of cooling.

We will leave this complicated subject with a single suggestion. If passage through martensite is, as we believe, a most efficient machine for damaging steel and producing trouble spots, and if rails as now cooled on the hot bed have to run this fell gauntlet, and if the molten salt bath quench escapes martensite entirely, why should we not produce a very superior rail with the molten salt bath quench? One of the major reasons why the rail mills have been so loath to develop heat treated rails has been the tremendous problem presented by the facilities necessary for tempering the rails adequately after quenching. If the contentions of the molten salt bath advocates are correct, why is it not entirely feasible to quench in the salt bath securing the desired structure directly from the quench and eliminating the tempering treatment entirely?

The oil well drill stem seems to stand almost in a class by itself. It operates utterly remote from observation, so that the stress cycle to which it is subjected can be judged only by the analysis of its failures. Unfortunately, the conditions are such that the evidence that might be gleaned from a close study of the fracture is largely destroyed at the time of observation. We are therefore forced to a regrettable degree to go by dead reckoning, in our attempts to improve the steel for this service.

Both the inner and outer surfaces of the tube constituting the stem are of utterly vital importance. In the absence of close knowledge of the stress cycles imposed on the stem, we will state once for all that we have less assurance of the accuracy of our qualitative analysis which follows than is the case in simpler problems.

We judge that the outer surface metal of the tube is subjected to the heaviest service stress of any of the metal in the stem. This stress is a combined torsion and bending, with both stresses varied by rapid, irregular and violent vibration, and with the further complication of abrasive wear.

The inner surface metal of the tube, as well as the entire body of metal inside of the outer surface metal is also subjected to much the same type of stress cycles, but with a little less stress intensity.

In some respects, however, the inner surface may be more critical than the outer surface. This is due first to the fact that in spite of all the care that is taken in the tube mills, the inner surface of tubes forty feet long cannot be inspected for damaging defects as efficiently and as surely as the outer surface. We say this without hesitation, and it must be admitted when we realize that it is not always possible to detect damaging weak streaks (as defined earlier in this paper) even on the outer surface of the tube, with any inspection short of a destructive deep etch.

The outer surface of the tube is made up of the outer longitudinal surface of the ingot very much as in the cases previously considered.

The inner surface of the tube is made up out of the segregated material lying in the core of the ingot, and this is another reason why the inner surface of the tube may be critical in its relation to service failures.

The H-section locomotive connecting rod is made in current practice by forging a body rectangle capable of being machined down to the finished H. This heavy body is heat treated (when a heat treated rod is specified) and then the H is machined out. Some makers forge a very slight approach toward the H but many rods are machined from the blank rectangle. This leaves the segregated core of the ingot nakedly exposed.

In freight service, the whole cross section of the rod is subjected to maximum alternating tension and compression stresses and there have been too many cases of the development of transverse fissures.

It is axiomatic to state that wherever a rod made by this process is satisfactory, a materially lighter rod could have been made by forging the H very close to finished size and then machining off just enough to remove the inferior surface material.

In fact it is entirely possible that a rod in which the H-section is forged to the finished size and not machined at all, would be not only better but also cheaper than a rod produced by the usual process.

### CONCLUSIONS

We urge the mechanical designer to learn enough of the steel making processes to know what part of the ingot goes into the critical location of every vital piece of his product, and to what extent

he must allow in his designs for the presence of weak spots in the steel in that critical region.

We urge the testing engineer to develop tests that will give us the knowledge we need as to the strength of weak spots in resisting fatigue.

We also urge the testing engineer to show us the effect of heat treatments on the surface strength of steel both machined and unmachined.

We urge the research metallurgical engineer to revise his methods of investigating transient length changes during quenching.

We urge the steel maker to bring the quality of the surface of the ingot of his greatly improved alloy and other steels closer to the average quality of those steels and to also eliminate surface decarburization from steel which is intended for heat treating and going into service without machining, and in this connection we again call to his attention the opportunity for building future profitable business by undertaking a comprehensive research covering the steel for some of the more difficult services falling within the classifications covered in this discussion.

## APPENDIX

### QUENCHING DILATOMETER FOR SALT BATH HEATED SPECIMEN

#### (A) Detailed Description

This description reads in part as though the dilatometer were fully developed into a practical machine, ready to work. This is not the case. The intent is to show that a practical instrument can be made but material further development of details is required.

The specimen (1) Fig. 1 is a thin-walled tube which is cylindrical throughout the gage length and is conically flared, outwardly at its upper end and inwardly at its lower end. The flaring at the ends is to facilitate the assembly of the grips which are equipped with heat insulation to minimize parasite heat flow to the specimen.

The lower end (2) of the specimen is gripped to a spider (3), which is shown as an integral part of the stem (4), by means of a shrink ring (5).

Insulations (6) and (7) retard heat flow from the ring and spider to the specimen.

The stem (4) is interrupted by an insulated joint (8) just above the specimen and above this contains two universal joints (9) and (10) for purposes of alignment of the extensometer parts. These universal joints are free from end play. They appear essential on account of the difficulty of assembling insulated joints in accurate alignment.

The upper end of the stem (11) is firmly connected to the armature (12) of the extensometer.



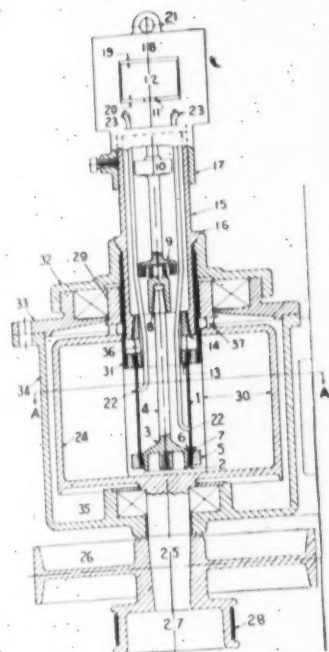


Fig. 1

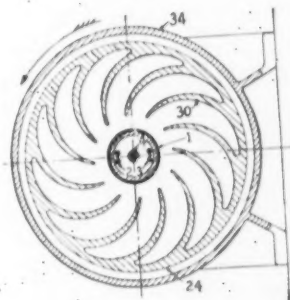


Fig. 2

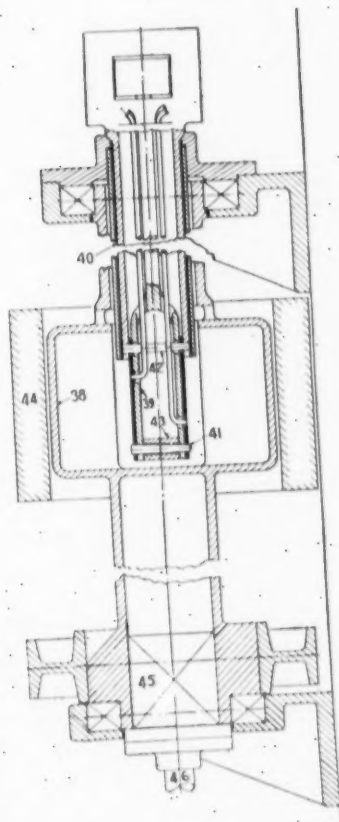


Fig. 3

The upper end of the specimen is similarly connected to a lanterned nipple (13). This nipple is connected by an insulated joint (14) to a tube (15).

(In regular use of the instrument it appears essential to have assembly jigs to facilitate the proper gripping of the specimen.) This tube is provided with a coned seat (16).

It is also interrupted by a length adjustment tube (17) to center the extensometer air gap accurately after the specimen grips are snugly home in place.

The upper end of the adjustment tube (17) is firmly connected to the frame (18) of the extensometer.

We will not go into a detailed description of the electrical connections of the extensometer. It is sufficient for our present purpose to state that the principles of the extensometer have been thoroughly developed by the Westinghouse Electric and Manufacturing Co. at East Pittsburgh.

Voltage is induced across the air gaps (19 and (20).

Movements of the armature relative to the frame produce a variation in this induced voltage on a straight line characteristic. Thus with proper calibration the induced voltage as read on the oscillograph is a direct measure of relative movement of armature and frame and therefore of the change of length of the specimen. Relatively small length changes will also occur in the structural parts of the extensometer. The various heat insulations are provided to minimize these errors. The relative masses of specimen and extensometer parts also reduce these errors.

An eye (21) is provided to facilitate handling of the assembled specimen and extensometer.

Thermocouples (22) are introduced from inside the specimen into holes drilled through the wall.

The terminals are then welded to the wall from the outside.

Later description brings out very clearly that a real problem may be presented by the insulation of the lower part of the thermocouple leads. These leads leave the extensometer at its upper end as shown at (23).

The quenching rig consists of a cylindrical tank (24) which is mounted on antifriction bearings on integral shaft extensions.

The lower shaft extension (25) carries a brake drum (26) and a driving pulley (27).

A belt (28) is the driving connection from an electric motor which is not shown. As an alternative arrangement, the motor may drive direct (See 46, Fig. 3) through a slip clutch, as shown at (45) of Fig. 3. The direction of spin is indicated by the arrow in Fig. 2.

The upper shaft extension (29) is hollow.

The inside of the tank is equipped with webs (30) which are of an involute or in general similar curved shape, as shown in Fig. 2. (A little license has been used and this view shows the quartz connection to the extensometer armature as shown in Fig. 2).

A tubular guide (31) extends through the opening in the hollow shaft (29). This guide is operatively integral with the bearing cap (32) which in turn is securely attached to the cover (33) of the stationary casing (34) which surrounds the quenching tank and forms the foundation of the instrument.

#### (B) Procedure

The assembled specimen and extensometer are picked up by the eye (21) and lowered into a bath of molten salt which is being held at the proper temperature.

The first lowering into the bath is stopped temporarily when the shrink ring (5) is just in contact with the surface of the bath. This is to permit the ring (5) and the spider (3) to come up together in temperature. The protrusion of spider (3) below ring (5) and the relative mass of (3) and (5) are arranged to tend toward this uniformity of temperature rise, which is aimed at preventing the loosening of ring (5).

It is clear that the materials of 3 and 5 must be selected with regard for this problem of ring tightness.

After the lapse of a time interval determined by experience the sample is

1929

immersed till the holes (36) in the lanterned nipple (13) are about half to two-thirds covered.

The molten salt runs into and fills the inside of the specimen.

The mass and material of the nipple (13) and the co-operating shrink ring are selected on the basis of the same fundamentals as 3 and 5.

The insulated joints (8) and (14) are for the purpose of retarding heat flow up into the extensometer parts.

The specimen is then raised till the upper surface of the upper shrink ring just emerges from the bath, and it is held in this position till the meters in the thermocouple circuit show equilibrium at the desired temperature.

Just before this equilibrium is attained, current is switched onto the quench tank spinning motor. Centrifugal action will then form an axial vortex with practically vertical walls in the quenchant.

The tank is so proportioned that the volume of the vortex is equal to the volume of the tank above the bottom of the guide (31) when in its position shown in Fig. 1. Any excess quenchant is then automatically spilled out at the top of the tank, passing out through holes (37) and down to the sump (35). In order to permit this spilling of excess quenchant, a narrow space is provided between the guide (31) and the opening in the top of the quenching tank.

The quenching rig is now ready to receive the specimen.

There is slack in the extensometer and thermocouple leads, sufficient to permit the following movements, which are carried out manually and as rapidly as possible.

The oscillograph is started.

The specimen and extensometer unit are lifted from the salt bath and lowered through the guide (31).

Current is snapped off from the spinning motor and onto the electric brake which acts on the drum (26).

The tank spin is abruptly reduced to a slow rotation, the motor connections being arranged with this end in view.

(The advantage of a slip clutch direct drive, referred to previously, over a belt drive lies in the danger of the belt jumping off during the sudden change of speed. To minimize this danger a very loose belt is used).

When the speed occurs, the kinetic energy stored in the quenchant, advances the quenchant against the involute webs which deflect it inwardly against the specimen.

It is obvious that the quench tank requires dynamic balance under operative conditions. This is quite a problem, but can be worked out.

#### INDUCTIVE HEATING VERSUS SALT BATH HEATING

Fig. 3 indicates a possible disposition (44) of an inductive heating coil. This method of heating the specimen has some very attractive features.

It eliminates the major part of the time element elapsing between the arrival of the specimen at the desired temperature and the actual contact of the quenchant with the sample.

It eliminates the complications of the contact of molten salt with the thermocouple leads, etc.

It eliminates the complication of the solidification of the contained salt before the sample reaches the temperature of the quenchant.

It eliminates the danger of the loosening of the extensometer grip on the specimen.

It involves radical modifications in the construction of the quenching rig.

All parts, other than the specimen, which lies in the inductive field, must be constructed of nonmetallic material.

This is quite simple except in connection with the quenching tank (38) Fig. 3. This must be made from a molded material such as a reinforced bakelite, for example. As it is shown in the Fig. it is not capable of being molded, the Fig. being only diagrammatic. It is technically possible to mold a structure which will fulfill the necessary functions. It would probably be a composite rather than an integral structure and the molding dies would be very expensive.

The extensometer armature connection (39) and frame connection (40) are preferably of fused quartz.

The specimen attachments are taper pins (41 and 42) of the same material.

The extreme bottom of the extensometer armature connection (39) is closed by a quartz plug (43).

The slight leakage of the quenchant to contact with the inside of the specimen at the ends introduces a negligible error.

An error also enters in connection with hot looseness between the pins (41 and 42) and the specimen. We have found no way to eliminate this type of error and secure an absolutely fixed gage length for the specimen.

We judge that specimens with wall thickness down to about 20 thousandths of an inch are quite practical. As lower thicknesses are attempted the cost of the specimens will increase.

The shrink rings of Fig. 1 are a difficult problem. We judge that nickel will be the best material for the extensometer connections.

The technique of thermocouple attachment will require very careful study, particularly in connection with salt bath heating.

## DISCUSSION

**Written Discussion:** By P. G. McVetty, Westinghouse Electric and Manufacturing Co., East Pittsburgh.

Mr. Eaton has given us an excellent discussion of a subject which is vital to the proper application of heat treated steels. From the point of view of the testing engineer, it is interesting to note the growing conviction that failures result from "regions of weakness" which existed in either the skin or the core of the original ingot, and which were not removed in any subsequent work or heat treatment. If the full significance of this one point can be impressed upon every designer, this paper will have rendered a valuable service.



The theory of elasticity and all the formulae used by the designer are based upon the assumption that the material is homogeneous and isotropic. It is not surprising that failures occur in service when this basic requirement may not be realized. It should be emphasized that the factor of

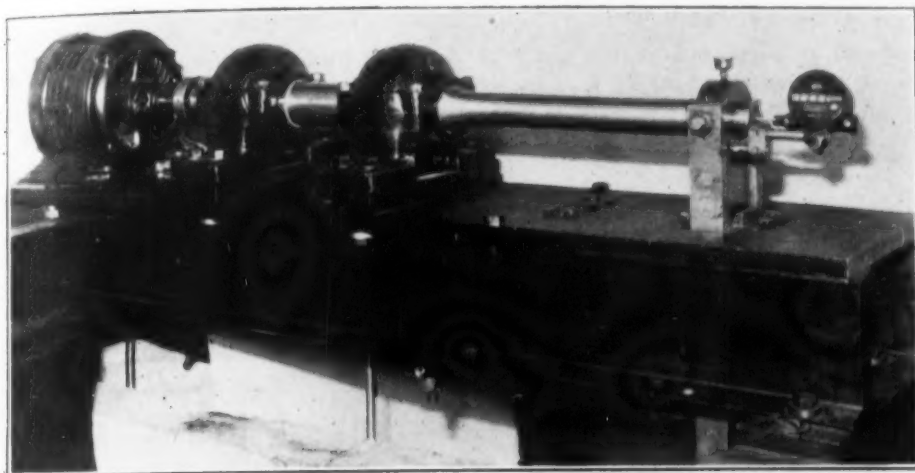


Fig. 1—Fatigue Testing Machine Using  $3\frac{1}{8}$ -inch Specimen.

safety must cover this lack of uniformity of material in addition to any other uncertainties of design. This is particularly important in applications involving repetition of stress or shock loading. We have been investigating this subject for several years and the results have been of great value to us.

Mr. Eaton appears to question the value of some of the fatigue testing work which has been done. A careful reading of his paper, however,



Fig. 2—Comparative Size of Various Fatigue Specimens: A—Cantilever,  $2\frac{1}{4}$ -inch; B—Cantilever, 1-inch; C—Cantilever (McAdam type), 0.4687-inch; D—Beam (Modified Farmer type), 0.3-inch; E—Direct Stress (Haigh type), 0.196-inch.

shows that his objection lies in the application of data obtained from machined specimens to conditions involving the use of unmachined material. The importance of developing fatigue tests in which the surface of the test specimen is truly representative of the critical section of the material in service cannot be overestimated. We are now working on some tests of this kind.

Mr. Eaton has mentioned fatigue tests of large specimens which were described by R. E. Peterson at the meeting of the American Society for Testing Materials last June. It may be of interest to show here a photograph which brings out the relative sizes of different test specimens.

Fig. 1 shows the large machine for which specimens are cut from bars at least  $3 \frac{3}{16}$  inches in diameter. Fig. 2 shows the comparative sizes of various fatigue test specimens. This machine runs at 1200 revolutions per minute and it will apply a stress of about 35,000 pounds per square inch to a critical section  $2\frac{1}{4}$  inches in diameter. In connection with a general study of fatigue phenomena Mr. Peterson has since developed a machine for small specimens 0.05 inch in diameter. He is investigating the effect of size of specimen and the results of these tests will show what, if anything, is to be gained by tests of still larger specimens.

In conclusion the writer agrees heartily with Mr. Eaton's suggestion that some steel maker conduct a comprehensive research not only to improve the surface of his product but also to improve its general uniformity. If the present uncertainty regarding the uniformity of steels can be removed, it will be possible to provide the designer with more reliable test data. It will be possible also to enlarge the scope of our tests to bring them closer to service conditions. Much development in this direction is now hampered by the difficulty of securing uniform material for test purposes.

**Written Discussion:** By Howard Scott, Westinghouse Electric and Manufacturing Co., East Pittsburgh.

We have Mr. Eaton to thank for a penetrating analysis of rather obscure phenomena of heat treatment which may influence the serviceability of steel under alternating loads. This analysis is particularly welcome at the present time when endurance testing is well established as a test method, but the study of factors influencing fatigue strength is hardly begun. Having had occasion to study this general subject with the same objective, but a different viewpoint, the writer arrived at a different evaluation of the factors involved. As the evaluation of the relative importance of the factors involved largely determines the choice of critical experiments, it is perhaps appropriate to discuss Mr. Eaton's propositions from the alternate viewpoint.

Mr. Eaton's main thesis is that drastic quenching is harmful to the endurance strength of steel. If this is universally true, much steel is being abused. On the other hand, if drastic quenching is harmless, there is considerable unnecessary waste in the use of alloys. The question introduced is, therefore, one which deserves the most searching inquiry.

It is quite improbable that a categorical answer to the question: is drastic quenching harmful to the endurance of steel, will be correct. Consideration of the negative side is, nevertheless, of interest because of the possibilities for improvement in structural alloy steels which it reveals. Drastic quenching may be defended on the basis of observation on the manufacture of ball bearings. Steel balls for bearing service are given the most drastic quench feasible. In spite of this treatment they have

extremely high resistance to fatigue failure. The advantages of drastic quenching to the ball manufacturers is obvious. No higher alloy content is used than is necessary to produce complete hardening in each size class. Consequently the ball manufacturers are in the enviable position of using alloying agents in steel with very high efficiency. This example raises the question: cannot the users of structural alloy steels achieve the same efficiency in use of alloying elements in steel?

This question cannot be lightly answered because of the complex shapes in which structural alloy steels must be treated for one thing. Of course many simple shapes such as leaf springs are used where drastic quenching may be feasible, but the complex shapes are the more important class. Neglecting for the moment the possibility of invisible damage from drastic quenching, visible damage in the form of cracks will occur in complex shapes if the carbon content is high as in bearing steels. To avoid this difficulty moderate carbon contents are required. The virtue of moderate carbon content lies in the fact that the steel as quenched has sufficient ductility to prevent visible cracking. It would appear, therefore, that such steels can be water-quenched with safety.

Mr. Eaton opposes the foregoing proposition on the grounds that drastic quenching causes sub-microscopic damage to the surface of the steel. Although one may not subscribe unconditionally to this opinion, it is nevertheless desirable to investigate means for avoiding the possibility. One means is to lower the carbon content further yet and thereby increase the ductility of the steel as quenched, but this requires an unacceptable sacrifice of strength. Another possible means is suggested by Mr. Eaton's statement that carbon steel is superior to alloy steel for some applications.

When carbon steel is compared with alloy steel on the basis of equal static strength, it is necessary that the carbon steel have an appreciably higher carbon content. This difference is quite significant as regards the relative content of nonmetallic inclusions. Normally the higher the carbon content of a steel, the lower will be its oxygen content and consequently the lower its content of nonmetallic inclusions. If these inclusions lower the endurance limit of steel materially, as is generally assumed, the higher carbon steels should have the better endurance limits for the same ultimate strength. This observation is particularly significant in connection with the endurance of low carbon alloy steels.

Structural alloy steels are much lower in carbon content than the bearing steels and are correspondingly more dirty. Presumably their endurance limit is lower than if they were equally clean. If so the formation of microscopic or sub-microscopic cracks on drastic quenching is probable in the presence of inclusions, but unlikely in their absence. Mr. Eaton would avoid this source of damage by hardening expedients, namely, air-hardening or quenching in a fused salt bath. The first mentioned expedient, however, is not practicable because of the additional cost of higher alloy content and decreased machinability. The second expedient does not avoid martensite formation except when the end product is sorbite and then the hardening is inadequate. In view of these objections

to modification of quenching practice, the possibility of eliminating inclusions to achieve the same end deserves consideration.

The inclusion content of steel may be substantially reduced by use of high carbon contents without change of steel refining practice. The use of a high carbon content in structural alloy steels would be feasible, however, only if a ductile martensite could be obtained with it. The possibility of developing a ductile martensite by alloying is certainly remote, but the advantages to be gained therefrom make the proposition worthy of study.

There still remains the possibility of eliminating inclusions by improved steel furnace practice. Here the urge for improvement is lacking because the presumed harmful effects of inclusions have never been experimentally established. The urge ought to be supplied by prompt experimental attack on this problem. An additional incentive for this action is offered by the prospect that great advantage may accrue simply from control of inclusion composition.

The outlook for a successful attack on the foregoing problem is encouraging because it is now possible to produce steels free from inclusions in sufficient quantities for mechanical testing. I refer to the vacuum melting furnace of Brace and Ziegler. With an inclusion-free steel for a datum point, the effects of inclusions can be definitely established. Herty and associates at the Bureau of Mines have shown how the quantity and composition of inclusions can be predetermined. Jordan and co-workers at the Bureau of Standards have contributed practicable analytical methods for inclusions in the finished steel. Hence the effect of inclusions on fatigue strength need no longer remain a subject for conjecture.

These remarks simply amplify somewhat the general impression created by Mr. Eaton, namely, that there exists a gross deficit of fundamental information on the treatment of structural alloy steels as related to composition. The correction of this situation to the extent of putting the use of alloying elements in steels for service under cyclic stresses on an efficient basis is a large and complex problem. There is little prospect for a successful solution in a reasonable time except by co-ordinated attack. If Mr. Eaton's paper inspires such an attack, it will have performed a highly commendable service.

**Written Discussion:** By W. J. Merten, Westinghouse Electric and Manufacturing Co., East Pittsburgh.

This discussion covers two phases of the paper:

1. Surface finishes;
2. Plastic flow phenomena.

We all recognize that outside of structural uniformity and selective grain refinement, surface finish is probably the most important factor contributing to the success or failure of a part during treatment as well as in service where the exploitation of the physical characteristics of steel becomes a practical necessity. It is here where our quantity production methods fail to keep step with metallurgical progress.

Refinement in grain structure and consequent subjection of these parts



to higher stresses automatically introduces greater sensitiveness to machine toolmarks, fillets and all surface imperfections.

Mr. Eaton's paper certainly is a timely one and whatever difference of opinion there may exist in the detailed procedure the endorsement of the handling of the general problem is certainly forthcoming. There is a question in my mind regarding the degree of perfection and suitability of a ground surface on springs, for example, especially when grinding is followed by open fire heating. When examining a ground surface under a microscope the improvement is not so apparent and still greater refinement suggests itself. Some experiments conducted with a ground and drawn bar, offered quite some promise, although unless subsequent heating is done so as to preserve the surface the expenditure is hardly worth while.

The other phase of Mr. Eaton's paper I want to say a few words about is plastic flow with respect to warping, failure of parts and internal stresses resulting during heating, quenching and tempering.

We all have analyzed effects of plastic flow during heating and quenching and failures due to localized plastic deformation resulting in internal stresses of uncertain magnitude. When this plastic flow phenomenon is applied to the relieving of internal stresses on tempering the exact location of the highly stressed section of the steel part is quite essential to avoid fracture during tempering, which operation in the light of this paper must also be a plastic flow process.

**Written Discussion:** By T. McLean Jasper, director of research, A. O. Smith Corporation, Milwaukee.

The author is to be complimented on bringing a new thought into the question of the possible effects of quenching on alloy steels. The idea being that transient weak streaks will develop on the surface of quenched material which should subsequently be removed by machining or grinding in order to produce a part free from the possibility of contributing to insipient failure.

The question of surface weakness may possibly be associated more than is usually done with the condition of the metal surfaces brought about by the method of handling it in the soaking pits and in the furnaces during heat treating operations.

Along the line of reducing the violence of quenching and also of providing a more economical method of heat treating, the writer some three years ago advocated, in the case where the heat treatment involved a quench and a subsequent tempering, that if steel were quenched in a hot bath that only one heating would be necessary. In other words, if a quench from above the critical point was to be performed and a temper of 900 degrees Fahr. were advocated that if the metal were quenched in a bath kept say at 800 degrees Fahr. it would acquire similar physical properties as a quench and subsequent temper of 900 degrees Fahr. At the same time the risk of outside cracking and overstressing in the quench to ordinary temperatures would be reduced. In the case of some alloy steels considerable trouble due to outside cracking has been experienced with too violent a quench.

The light which Mr. Eaton brings on the transition temperatures of alloy steels is a valuable contribution to the general thought. The recognition of buried weakness brings us to a problem of working for cleaner steels. The writer believes that this will be an accomplishment of the next few years because of the fact that the value of cleanness is being appreciated.

There is one other important matter touched upon by the author, and that is the matter of surface decarburization of steel during heating in the furnace. This has been found to be a very important factor even in carbon steels. Associated with decarburization is overheating with its attendant brittleness. Furnace atmospheres even in the soaking pits before the steel is rolled should be carefully studied because decarburization is almost eliminated in a deoxidizing atmosphere. Subsequent heat treatment should also care for this important factor.

I am wondering, therefore, if the question of the necessity for surface removal in an alloy heat treated sample is not closely associated with the care which the steel has had during rolling and subsequent heat treatment.

It will be noticed that the writer has avoided a discussion of this paper from the metallurgical standpoint. This has been deliberately done so as not to confuse what may be due to other conditions with what may be considered entirely metallurgical.

**Written Discussion:** By H. C. Loudonbeck, chief chemist, Westinghouse Air Brake Co., Wilmerding, Pa.

We have read Mr. Eaton's paper with considerable interest, and we feel that it is well worth while considering; that is, the consideration should be in the way of making some actual tests and experiments along the line which he suggests. It is rather difficult to discuss a paper of this character without some evidence confirming or showing that he is wrong in his conclusions. There is no evidence as far as I can determine of setting up severe strains in the quenching operation, providing the piece is of proper shape and is designed in the proper way; that is, all the strains from the quenching operation should be removed by suitable tempering, after the quenching operation. We have always maintained that the tempering operation should be sufficient to take care of this.

If a specification is followed which allows little tempering to meet the proper physical properties, naturally more or less strain will be maintained in the finished product, which is detrimental to the service of the material. However, we maintain that Mr. Eaton's suggestion should be followed up with experimental work for we feel that this is absolutely necessary before the merits of the paper can be determined; in other words, the Baconic principle of truth should be applied.

**Written Discussion:** By J. R. Freeman, Jr., senior metallurgist, U. S. Bureau of Standards, Washington, D. C.

This very thought-provoking and timely paper by Mr. Eaton presents in a clear and concise manner his conception of dimensional changes and resultant stresses that must occur in a piece of steel during cooling. It is

admittedly speculative and to use his own expression many will probably 'rise to the bait.' There are one or two points that I would like to discuss. Mr. Eaton uses the term, martensitic dilatation. This terminology appears rather unfortunate. By it, he refers, of course, to the dilatation that occurs when gamma iron changes to alpha iron on cooling. This dilatation occurs, so far as is known, in pure iron, at least in a practically carbon-free iron where certainly no martensite forms on cooling. On page 827, paragraph 2, it is stated that due to the fact that carbon is absent, little or no martensite can form and he therefore apparently assumed on page 828, paragraph 5, that no dilatation will occur in the relatively carbon-free surface layer during cooling. The fact that a marked expansion may occur in this decarburized iron is apparently overlooked unless he is assuming that the quenching temperature is below the  $A_3$  transformation, the temperature of which will depend upon the completeness of the decarburization. On page 836, it is stated that a rail steel on cooling on the hot bed passes through the martensitic structure. This is certainly open to question and depends upon the theory of the decomposition of austenite to which one subscribes. In the very clear picture that he gives of the causes of the deflection cycle followed by a rail on cooling, it is not necessary to assume the formation of martensite. The same relative contractions and expansions would occur in a rail made of relatively pure iron due to dilation occurring at the  $A_3$  transformation. The reasoning, therefore, that a salt bath quench by eliminating the formation of martensite would eliminate the deflection wave, does not hold.

On page 841, it is suggested that a program of fatigue tests should be inaugurated in which the specimens should be tested in their 'as forged,' 'as rolled' or 'as heat treated' conditions. Isn't this putting the cart before the horse? Would it not be a more logical procedure to make a systematic study of the effect of known reproducible surface conditions. Mr. Eaton is really seeking information on the 'notch effect' in fatigue. Once the effect of certain determinable notch conditions is determined, it should be possible to correlate them with the 'as processed' conditions found in practice. This question of notch effect in fatigue in its broadest sense includes the field of the effect of metallic or other coatings on fatigue properties which Mr. Eaton should be pleased to hear we are investigating at the present time.

#### Oral Discussion

F. H. CLARK: On page 842 the author speaks of fatigue failures due to the reopening of ingot ruptures which have been closed and perhaps partially welded during the working of the steel. I would like to ask if there is any direct evidence for this statement such as study of the fracture or microstructure.

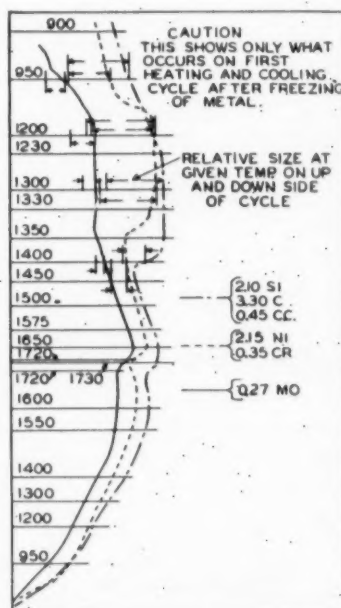
G. M. EATON: I am simply going to pass the buck on that. I was afraid when I wrote this paper that I would hurt the feelings of some of the steel manufacturers in condemning the surface of the improved production, and to check it up I sent the paper to Earl C. Smith of Central Alloy Steel Corp. and asked him about it. He said the paper is open to very severe criticism

because it understated the situation; in particular he mentioned this feature, among other things, which Miss Clark refers to. So I will just pass the buck to Earl Smith. I do not know.

I would like, Mr. Scott, to call attention to the fact that the illustrations which you used as proving that drastic quenching does not produce the effects the paper suggests unfortunately does not apply. It is stated 3 or 4 times in the paper that discussion applies to steel which is in the forged condition. Now, in the ball bearing, if you can imagine a set of bearing balls which are made in that way, forged to the finished shape, quenched, and put to work, I would like to entertain a little bet with you that you will get utterly different results from what you get with the ball bearing where after the surface has, if we are right, been damaged. That damaged material is all taken off, cracks and all are cut out and your illustration absolutely does not apply. Otherwise, I agree with your discussion and I want to thank you for the thought you put on it.

#### Author's Closure

Mr. Freeman's discussion is very constructive. In the brief time available before turning in our reply, we have secured enough comment on



what Mr. Freeman states to lead us to suspect that we must accept his criticism of the term "martensitic dilatation."

As soon as we make this admission, we have automatically undermined the value of many passages in the paper if they are judged on their exact wording. The underlying thought of the paper, however, remains unchanged. In fact, the thought Mr. Freeman's discussion inspires makes it possible to define the critical dilatation characteristic in a more tangible manner. The accompanying curves show condensed actual dilatation



curves of three types of grey iron. We have called these actual dilatation curves. This means only that they were made in a commercial type of dilatometer. We will refer to this in more detail a little later.

It will be noted that there is a material difference in the slope of the three curves as they pass through the critical points. This means that the dilatation curve for a given material (if accurate) gives some relative measure of the severity of local plastic flow which occurs when this material has inflicted upon it the steep temperature gradient involved in quenching. Before we can accept the indications of any dilatation curve however we must be convinced that it gives a true precision picture of the actual dilatation throughout the critical range.

The curves in the figure unfortunately cannot possibly be accepted for this purpose and, they are given solely for the purpose of illustrating the point. These curves were developed with a temperature cycle which was slowly but continually changing. The samples were around  $\frac{1}{2}$  inch in diameter and very short. Therefore the lengths as measured were at all times the result of an average of conditions which varied both axially and radially. In spite of this fact the slope was very abrupt at 1400 degrees Fahr. in the unalloyed iron. If the actual dilatation of a minute unit mass could be measured it would be found to be much more abrupt.

This situation calls attention rather more forcibly than does the paper to the desirability of the development of means for producing dilatation curves which approximate the actual more closely than can now be done even by the best step by step cycles and small specimens now employed.

We believe dilatation curves which are nearly correct will give a good index of the severity of quench that can properly be given to various steels. This applies particularly to articles which have the mill surface in an unmachined condition.

There is another fundamental characteristic of any given steel which is involved in judging the ability of the steel to resist drastic quenching. This is the ductility of the steel at the critical temperature of maximum dilatation. As so often happens there is a chance of being misled by the information which would be available if we knew the precise rate of maximum dilatation as secured from a very thin test piece heated inductively with small temperature gradient, and if we also knew the elongation and reduction of area at the critical temperature. The point at issue is covered in the paper, viz: does plastic flow under tension assist the transformation during which an increase of specific volume occurs? If it does we may not be justified in predicting the reaction of a heavy mass from the action of a very small mass.

It seems to us that the paper, as assisted by Mr. Freeman's contribution opens an attractive possibility of developing tests for evaluating the quenchability of various types of unmachined steel.

## SELECTION OF CASE HARDENING STEELS FOR HIGHLY STRESSED GEARS

BY H. W. McQUAID AND O. W. McMULLAN

### *Abstract*

*This paper describes the results of experiments made in investigating the most desirable steels for carburized gears. In Section I is set forth some of the desirable qualities that a gear steel must have and the various factors influencing these qualities.*

*Section II takes up the experimental work done. Surface hardness tests, depth-hardness curves, transverse loading of bars, notched and unnotched impact results, and dimension changes on bars and test rings are recorded. The steels tested include several of the more common gear steels and a few new steels which are now commercially available. A few general comparisons are made as to which is the most desirable steel to use under a given set of conditions, and it is claimed that cyanide bath heating is not an advantage in obtaining high surface hardness on case hardened steels.*

### COMPARATIVE PROPERTIES OF COMMERCIAL CASE HARDENING STEELS AFTER CASE HARDENING

**I**T IS probably true that very few mechanisms are designed to operate under such high stresses as are the gear teeth in the average automotive rear axle and transmission. In addition to having to carry an extremely high load, the gear teeth must have a maximum resistance to wear and hence the selection of a material for this purpose is of the utmost importance as far as the service is concerned.

The manufacture of ring gears and pinions to satisfy the requirements of strength, resistance to wear, and quietness is one of the most difficult problems which the axle builder has to contend with. The quietness is a function of the surface condition of the teeth, the design, and the distortion in heat treatment. The resistance to wear depends upon the hardness of the surface, the structure of the contact surface, and the character of the bearing, which in

---

A paper presented before the Eleventh Annual Convention of the Society, Cleveland, September 9 to 13, 1929. The authors, members of the society, are associated with the Timken-Detroit Axle Co., Detroit. Manuscript received July 15, 1929.

turn is dependent upon the design, distortion, etc. The strength of the tooth is a function of the material, the heat treatment, and the character of the surface, and the design.

It is evident that the selection of a steel which will machine smoothly, distort little in heat treatment, and which at the same time will have the strength and toughness necessary to withstand suddenly applied loads and high ultimate stresses requires considerable experience and judgment. In addition to the above factors, the question of cost is generally of primary importance, although in this respect the factor of material cost is often confused with the factor of final cost.

#### FACTORS AFFECTING REQUIREMENTS OF STEELS FOR CASE HARDENED GEARS

The steels used in automotive gearing can be classified into two distinct classes. The high carbon oil-treated gear, and the low carbon case hardened gear. In transmission work, the high carbon oil-hardened gear is confined principally to the passenger car and light truck field. In truck and bus transmissions which are subjected to very heavy tooth load and considerable abuse, practically all gears are made from carburized steel and the use of the higher carbon steels for this use has been found to be uneconomical, because of the unsatisfactory results in service. In rear axle applications the choice of the low carbon carburizing steels is practically unanimous. It is safe to say that with one or two rare exceptions only carburizing steels are used in ring gear and pinion work.

The advantages of the low carbon carburizing steels are that they permit of easy machining and it is easy to obtain an extremely hard surface to resist wear, combined with excellent resistance to suddenly applied loads. The disadvantages of the low carbon carburizing steels are that they are apt to distort considerably in treatment, and the treatment itself is expensive and sometimes difficult to control.

It is a well known fact that the case hardened steels have a higher endurance limit than higher carbon oil-hardened steels, which would be satisfactory as far as hardness is concerned. Results have shown that case hardened fatigue tests practically always show a higher endurance value than corresponding higher carbon oil hardening steels, and it is for this reason that the case hardened gear is superior.

Plain carbon steels such as S.A.E. 1020 are practically never used in automotive work for ring gear and pinion applications, because of their lower strength and toughness and the greater tendency to warp in the heat treat. The common steels which are used for the purpose are S.A.E. 2315, 4615, and 6115. For especially severe applications S.A.E. 2512 and the so-called Krupp analysis is often used. Newer steels which have been introduced recently to compete with the above steels are the manganese steels, the manganese-molybdenum steels, the 3.50 per cent nickel-molybdenum steels, and in Europe the nitriding steels.

In addition to the analysis, there are many important factors which tend to affect the properties of the finished gear, and which are often of really more importance than the chemical specifications. There are several factors which affect greatly the relative distortion, the machining properties, and the toughness of case hardening gear steel. The method of manufacture is important, or rather it would probably be more correct to say the making of the steel is important. The grain size as indicated by the carburizing test is of the utmost importance. The cleanliness, the rolling temperature, and the method of forging all must be standardized if uniform results in practice are to be obtained. For continuous uniformity, and control of distortion, the electric furnace steels seem to give better results than the open-hearth steels. It is our experience that the variation in properties of one part of an open-hearth heat to another is greater than the variation from one electric furnace heat to another. This is particularly true where an effort is made to control the grain size to obtain uniform results in machining and distortion. Tests made by running an open-hearth steel against electric furnace steels of the same analysis indicate that it is impossible to control or predict the change in bearing with the open-hearth steel, whereas this has been found to be quite feasible commercially with the electric furnace steel. Inasmuch as the difference in price between the electric furnace steel as made by the large producers, and the better grades of open-hearth steel is relatively small, it will be found that the slight difference in price is not enough to offset the advantages of the uniformity of the electric furnace steel.

It has been found that the grain sizes as indicated by the carburizing test is one of the most important factors if uniform results are to be obtained in generating a smooth tooth surface. It has been



found that once the generating department has been set for a given grain size, that any radical change in the grain size results in an immediate upheaval in the generating department. Thus, if a grain size has been held to a range of say three to six as measured on the old grain size chart as adopted by the United Alloy Steel Co., that a change to 7 to 10 will immediately result in the gear division being at their wit's end to obtain a satisfactory finish. Hence it is important that the grain size be determined for the given analysis which

Table I  
Analysis of Steels

Steel No.	C	Mn	P	S	Ni	Cr	V	Mo	Si	Grain Size	Source
2	0.36	0.79	0.022	0.021	1.84	0.73	...	0.34	0.23	7-8	Electric Furnace
3	0.16	0.52	0.018	0.028	3.32	...	...	...	0.013	6-7	Open-Hearth
3A	0.15	0.45	0.018	0.028	3.42	...	...	...	0.099	7-8	Open-Hearth
4	0.13	0.51	0.015	0.021	4.99	...	...	...	0.16	10	Electric Furnace
4A	0.15	0.44	0.015	0.028	5.10	...	...	...	0.26	10	Electric Furnace
5	0.15	0.51	0.016	0.023	1.23	0.68	...	...	0.26	10	Electric Furnace
6	0.14	0.52	0.017	0.022	0.48	0.53	0.19	...	0.22	7-10	Open-Hearth
6A	0.12	0.46	0.015	0.024	0.53	0.62	0.18	...	0.17	10	Open-Hearth
7	0.13	0.54	0.015	0.044	1.75	...	...	0.25	0.23	10	Electric Furnace
7A	0.18	0.54	0.015	0.020	1.74	...	...	0.29	0.16	10	Electric Furnace
7B	0.15	0.58	0.018	0.012	1.82	...	...	0.26	0.12	10	Electric Furnace
7D	0.18	0.53	0.020	0.030	1.83	...	...	0.23	0.23	9-10	Electric Furnace
9	0.11	3.17	0.015	0.019	...	...	...	0.32	0.07	8-9	Electric Furnace
10	0.16	0.47	0.015	0.017	3.44	...	...	0.34	0.089	8-9	Electric Furnace

gives the best results with the tooling, etc., of the generators as established, and to insist on this grain size as long as no change is made in the gear department.

The analysis specifications are generally close enough to insure uniformity from this angle, and it is seldom found that much trouble is met with due to variation in chemical specifications, except in the higher alloy steels such as the 5 per cent nickel, where a narrow carbon range is very necessary.

In the manufacture of ring gears for applications where quietness and good bearings are essential, it is almost necessary to insist that the ring gear forgings be made by the up-set method. That is to say, the ring gear forgings should be made by up-setting bar stock to give the required "pancake" from which the ring gear is forged. The saving in price by using flat stock for forging ring gears is almost always absorbed many times by difficulties in the lapping department; and it is almost certain that if accurate checks are made, that the up-set ring gear forging will always prove to be an economy over the gear forged from flat stock. Up-set forgings show a mini-

mum tendency to distort, whereas the forgings made from flat stock are almost inevitably a source of "out of round" noise rejection.

The machining characteristics are an important factor in selecting a steel for case hardened gears, although many times one steel is condemned because it does not produce equal results in the machining operation when tried out under the same machine set-up as another steel which was previously tested. It is important to remember that co-operation on the part of the generator operator in adjusting feeds, speeds, etc., will often result in a condemned steel being found equal to, if not better, than the steel which was being used.

#### NORMALIZING

In the treatment of case hardening steels for gears, the normalizing is a matter of much greater importance than with any other forgings. It is generally desirable for the best conditions in the generating of the gear tooth that the forging after normalizing is not too soft. It is also important that the structure be a coarse grain pearlite and these conditions are, of course, influenced greatly by the finishing temperature in forging and the normalizing temperature. If the forging temperature is too low, and the normalizing also low, the structure after normalizing will be difficult to machine and the surface of the generated tooth not as good as it should be. For the best conditions on the generator, the finishing temperature in the forge shop should be high and the normalizing so regulated as to give a gear which is relatively hard. Most of the complaint in the beveled gear generator operation is due to the tearing of the surface which is produced by too soft and too tough a material. Much better results are obtained if the forgings are coarse-grained and hard. By carrying the normalizing operation at too slow a rate, the ordinary alloy case hardening steel such as S.A.E. 2315 will show a Brinell hardness of 140 to 149, whereas much better results are obtained in generating by regulating the normalizing so that the Brinell range will be from 156 to 179. It must be noted here that the Brinell hardness and structure which give the best results are dependent upon the feeds and speeds, etc., of the Gleason generator, and the best results can only be obtained by adjusting the normalizing to suit the Gleason set-up. It has been noticed that the ball bearing generator can take care of a much wider range in Brinell hardness, and still produce an excellent surface as compared to the

plain bearing generator. The generator with ball bearing spindle and work head are much freer from chatter and will no doubt replace the old style bearing very rapidly where a good surface on the ring gear and pinion is required.

#### CARBURIZING

As practically all ring gears and pinions are carburized, this process is of considerable importance to the gear maker. Most of the smaller production plants carburize gears and pinions in the same manner as they did twenty years ago, with the exception that alloy pots have most generally replaced the steel carburizing pot. For large production the continuous carburizing furnace is being quite generally adopted for gears, although its use is restricted to large production units.

The selection of carburizing compound and the economical practice generally followed out is quite uniform throughout the gear industry. It has been the experience of the writers that there is little choice, as far as the actual carburizing results that are obtained, between most of the compounds on the market, although the economy varies considerably from one compound to another, depending upon the stability of the compound and particularly of the binder used. Some compounds can be used with six parts of old to one part of new, while others require at least one part of new compound to each two parts of old. Since all compounds depend upon some form of charcoal together with sodium or barium carbonate as the primary elements, that compound should be selected which will result in the lowest cost per carburizing pot hour, which is primarily dependent upon the ratio of mixture, and the initial cost per cubic foot. It is usual to carburize gears so that the resultant case depth will be from 0.030 to 0.050-inch, depending upon the size of the gear and the use to which it is put. It is desirable to have the minimum amount of excess cementite in the case since the heat treatment generally given ring gears is not high enough to insure solution of more than a very small amount of excess cementite. For this reason the carburizing temperature and practice should be designed to keep the carbon content as low as possible. This is difficult to control, however, and we find most of the carburizing ring gear and pinion steels carrying a high carbon case. The temperature of the carburizing is generally regulated to suit the analysis, and varies generally from

1625 degrees Fahr. (885 degrees Cent.) for some of the high nickel steels, to 1700 degrees Fahr. (925 degrees Cent.) for some of the nickel-molybdenum and chromium-vanadium steels. The carburizing time is dependent to some extent upon the steel selected, and it is a factor which might be considered when deciding upon the steel for a given application.

### HEAT TREATMENT

The most common method of heat treating ring gears is to let them cool in the carburizing pot, reheat to a temperature above the

Table II  
Heat Treatment of Samples Tested

Steel No.	Treat-ment No.	Nor-malized Deg. Fahr.	Carburized Deg. Fahr.	Cool from Carburizing	Quenched Deg. Fahr.	Tempered Deg. Fahr.
2	1	1650	1675 12 hrs.	Air	None	400 45 min.
2	2	1650	1675 12 hrs.	Air	1459 Air Blast	400 45 min.
3	1	1650	1675 12 hrs.	Pot	1475 Oil	300 15 min.
3A	1	1650	1675 12 hrs.	Pot	1475 Oil	300 15 min.
4	1	1650	1675 14 hrs.	Pot	1450 Oil	300 15 min.
4A	1	1650	1675 14 hrs.	Pot	1450 Oil	300 15 min.
5	1	1650	1675 12 hrs.	Pot	1475 Oil	350 15 min.
6	1	1650	1675 12 hrs.	Pot	1600 Oil	350 30 min.
6A	1	1650	1675 12 hrs.	Oil	.....	350 30 min.
6A	2	1650	1675 12 hrs.	Pot	1450 Oil	350 30 min.
7	1	1650	1675 12 hrs.	Pot	1525 Oil	300 15 min.
7	2	1650	1675 12 hrs.	Oil	.....	300 15 min.
7A	1	1650	1675 12 hrs.	Pot	1525 Oil	300 15 min.
7B	1	1650	1675 12 hrs.	Pot	1525 Oil	300 15 min.
7D	1	1650	1675 12 hrs.	Pot	1525 Oil	300 15 min.
9	1	1650	1675 12 hrs.	Air	.....	350 30 min.
9	2	1650	1675 12 hrs.	Pot	1380 Oil	350 30 min.
9	3	1650	1675 12 hrs.	Pot	1450 Oil	350 30 min.
10	1	1650	1675 12 hrs.	Pot	1450 Oil	300 15 min.
10	2	1650	1675 12 hrs.	Air	.....	.....
10	3	1650	1675 12 hrs.	Air	1450 Air	.....

critical temperature of the case and quench in oil, using a quenching press. As a rule the quenching temperature is high enough to also "catch" the core so that the fracture from this practice is satisfactory. Pinions are as a rule cooled in the pot and reheated to above the critical point of the core and quenched in oil. Both ring gears and pinions are tempered at approximately 300 degrees Fahr. for a short time. It is a custom in some plants to quench directly from the carburizing pots and with some steels this practice gives the most excellent results. In many plants it is a custom to reheat pinions in a salt bath containing cyanide, and to quench in oil. It is claimed that the reheating in the cyanide bath results in greater hardness and cleanliness. For the higher alloy steels such as the 5 per cent nickel



steels, the heat treatment is usually the same as for the lower alloy steels. The heat treatment has a very important effect on the distortion characteristics, and hence every phase of it should be carefully determined and controlled with the possible effect on the bearing always in mind.

It is a well known fact that the grain size as determined by the carburizing test is an important factor in the proper quenching range

Table III  
Treatment Given Samples Heated in Cyanide Bath

Steel No.	Treatment No.	Nor-malized Deg. Fahr.	Car-burized Deg. Fahr.	Cool From Pot Carburizing	Cyanide Treatment Deg. Fahr.	Quench	Tempered Deg. Fahr.
7	C-1	1650	1675	Pot	1410 10 min.	Oil	300 15 min.
7	C-2	1650	1675	Pot	1410 30 min.	Oil	300 15 min.
7	C-3	1650	1675	Pot	1510 5 min.	Oil	300 15 min.
7	C-4	1650	1675	Pot	1510 30 min.	Oil	300 15 min.
7	C-5	1650	1675	Pot	1410 10 min.	Water	300 15 min.

and also an important factor in the core hardness obtained so that grain size must be considered in comparing two steels. The depth of case and the design of the dies in the quenching machine are also of great importance and some thought must be given to designing the dies so that warpage in the quench is reduced to a minimum. If these dies are properly designed, ring gears should come from the quench not more than 0.003-inch out of flat, or 0.003-inch out of round maximum, and at least 80 per cent of the gears should not show more than 0.0015-inch out of flat or 0.002-inch out of round.

It is a custom in most shops to run test gears and pinions from each lot machined or from each heat, to determine the change in bearing which occurs in carburizing and heat treatment, and to base the generating of the rest of the heat or lot according to the results obtained on the test lots. This is to be recommended, especially where the variation is great, and it is of value even under the best conditions to catch slip-ups which are bound to occur from time to time.

#### DETERMINING RELATIVE CHARACTERISTICS OF CASE HARDENING GEAR STEELS

The processing of ring gear and pinion steels to insure the best results in the finished gear is a subject which would require a paper

in itself to cover properly. The object of this paper is to set forth, if possible, the fundamental differences which are found in the usual types of case hardening steels for gears and also to determine the physical characteristics of some of the newer steels. The treatment of the steels for best machining and the determination of the factors which affect the machining have been purposely omitted as this is also a subject on which a long paper could be written. We have tried to compare the results obtained in heat treating more or less uniform sections to determine the relative tendency to distortion in the carburizing and quenching, and also the relative strength and impact values of the case of case hardened pieces. Although the greatest effort was made to obtain uniform conditions as regards case depth, etc., there exists some variation. The data obtained is indicative, however, of the relative values of the various steels under repeated stresses and shock loads. The change in bearing due to design, etc., we have assumed should be comparative to the change in section obtained on uniform bodies of the shapes used. Samples of nitriding steels which were also included in the tests gave good results comparatively, but we have not included them in the tables because of erroneous conclusions which might be drawn from the data given. We have found that the nitrided samples tested compare favorably with steels such as the S.A.E. 2315 and S.A.E. 6115. Results obtained with nitriding are, however, dependent upon the heat treatment and nitriding conditions. It is our opinion that the results obtained with nitriding steels at the present time represent only temporary values which, while comparatively excellent, will be greatly improved as our knowledge of nitriding increases.

## Section II

### METHOD OF PROCEDURE

*Carburizing.* All carburizing was done in the heat treat department along with production work. It was desired to obtain approximately the same case depth on all samples but as the work was not all done at the same time, a probable variation in conditions along with variations in analyses has caused some variation in the case depths obtained. In making this investigation, an attempt was made to hold closely to accepted commercial treatments for the different steel specifications. The depth of case is somewhat heavier

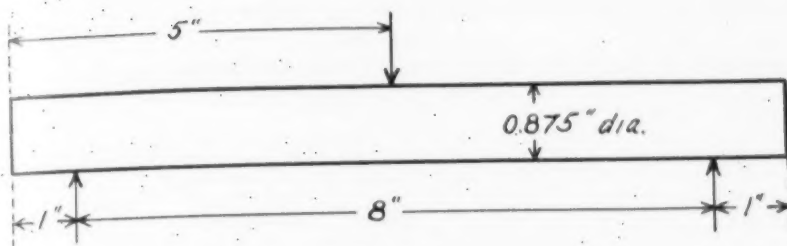


Fig. 1—Diagram Showing Dimensions and Methods of Loading Bars for Transverse Tests.

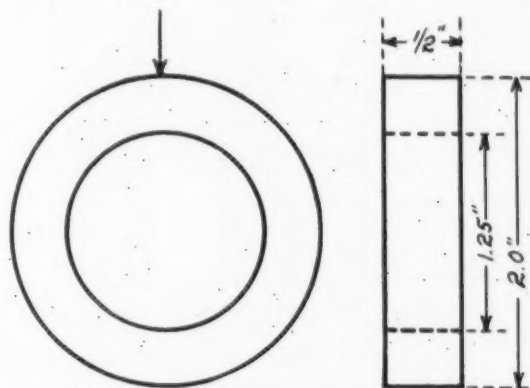


Fig. 2—Diagram Showing Dimensions and Method of Applying Load on Test Rings.

than it is usual to find in automotive work, but this is due to the fact that the writers are primarily interested in the heavier gear drives used in bus and truck applications. The case depth measurements given in Table IV were made on polished and etched sections with an optical micrometer. The pot-cooled samples were additional test pieces carburized in the pot along with the physical test bars. By saturated case is meant the total depth of eutectoid composition or higher carbon content, thus including the hypereutectoid zone. Total case depth is all that above a noticeable increase in carbon content over that of the core. The hardened case depth measurements given in Tables IV and V were made from the broken test specimens and measured to that depth, showing ferrite, troostite, or other constituent not common to the fully hardened higher carbon part of the case. These measurements were difficult to make, especially on the higher alloy or carbon samples showing a quenched core. The curves showing the hardness results on the taper-ground specimens give a better indication of hardened case depth. Those results were obtained on the impact plugs; but all bars, impact plugs, and rings bearing the

**Table IV**  
**Showing Case Depth of Pot-Cooled and Treated Samples and Surface Hardness of Treated Bars for Transverse Test**

Steel No.	Treatment No.	Case Depth on Pot-Cooled Samples			Hardened Case Depth on Test Bar	Hardness	
		Hyper	Saturated	Total		Rockwell on Case	Brinell on Core
2	1	0.018"	0.036"	0.063"	0.050" (?)	52 - 56	430
2	2	0.018"	0.036"	0.063"	0.050" (?)	60	512
3	1	0.013"	0.030"	0.048"	.....	62 - 63	321
4	1	.....	0.020"	0.035"	.....	61 - 62.5	402
5	1	0.010"	0.030"	0.043"	.....	62.5 - 64.5	241
6	1	0.009"	0.027"	0.045"	0.033"	42 - 57	196
6A	1	*0.012"	0.032"	0.062"	0.055"	59 - 61.5	223
6A	2	0.013"	0.032"	0.062"	0.060"	39 - 60	179
7	1	0.011"	0.028"	0.045"	.....	61 - 65	255
7	2	0.010"	0.021"	0.043"	0.050"	60 - 63.5	235
7	C-1	0.010"	0.021"	0.043"	0.038"	62 - 64	223
7	C-2	0.010"	0.021"	0.043"	0.043"	62.5 - 64	228
7	C-3	0.010"	0.021"	0.043"	0.054"	62 - 64	235
7	C-4	0.010"	0.021"	0.043"	0.052"	60.5 - 63	255
7	C-5	0.010"	0.021"	0.043"	0.045"	63 - 64.5	241
7A	1	0.018"	0.036"	0.060"	0.057"	61 - 63	262
7D	1	0.010"	0.024"	0.048"	0.050"	61.5 - 63.5	293
9	1	**	**	0.052"	0.048"	56 - 60	286
9	2	**	**	0.052"	0.045"	57 - 61	293
9	3	**	**	0.052"	0.055"	55 - 58.5	311
10	1	0.017"	0.035"	0.055"	0.045"	58.5 - 61	311
10	2	0.017"	0.035"	0.055"	0.040"	54 - 58	235
10	3	0.017"	0.035"	0.055"	0.040"	50 - 56.5	228

\*After recarburizing five hours.

\*\*These samples contained martensite in the case even when cooled in the pot.

For a comparison of air hardening properties it was found that carburized S.A.E. 2512 of similar size and treatment to the above bars gave a case hardness of 40-48 Rockwell.

same steel and treatment number were carburized at the same time and should have the same case depth, except for variation in size of the samples.

**Heat Treatment.** With the exception of the samples cooled in air or quenched from the pot and those treated in cyanide, all samples were heated in an automatically controlled electric furnace in the tool hardening department. Tempering was done in an oil bath. The cyanide bath used was 96 per cent cyanide except for decomposition through use.

**Dimension Measurements.** All measurements were carefully made by a well trained inspector and were read direct to four decimal places. The figures given in Table IX are from three readings on each of three bars and are the averages of the nine readings. The ring dimensions in Table X are averages on two samples. The samples were carefully cleaned from adhering foreign matter after carburizing and from scale after heat treating before measurements were taken.



**Table V**  
**Case Depth and Rockwell Hardness (surface) of Izod Impact Test Pieces and Transverse Ring Tests**

Steel No.	Treat-ment No.	Izod Bars		Rings		Remarks
		Rockwell on Case	Case Depth	Rock- on Case	Hardened Case Depth	
2	1	53 -57	0.055"	50 -54	0.055"	Cooled in still air from 1450 degrees Fahr.
2	2A	55 -61	0.055"	51.5-56.5	0.050"	
3	1	59 -61	0.045"	61 -62	.....	
4	1	58 -62	0.040"	61 -62.5	.....	
5	1	62 -64	0.045"	63 -64.5	.....	
6	1	53.5-61	0.045"	58 -63.5	.....	
7	1	61.5-63.5	0.045"	62 -64	.....	9-1 reheated 1400 degrees Fahr. Air-Cooled
7	C-1	63 -64.5	0.045"	.....	.....	
7	C-2	62.5-63.5	0.045"	.....	.....	
7	C-3	63	0.050"	.....	.....	
7	C-4	61.5-62	0.050"	.....	.....	
7	C-5	64	0.047"	.....	.....	
7A	1	61 -63	0.050"	.....	.....	
7D	1	62 -63	0.050"	.....	.....	
9	1	56 -63	0.052"	.....	.....	
9	2	58.5-62.5	0.050"	.....	.....	
9	1A	55 -62	0.050"	.....	.....	
10	1	60 -61	0.050"	.....	.....	
10	2	57.5-61	0.043"	.....	.....	
10	3	58.5-62	0.042"	.....	.....	

Case depth and hardness on Izod bars and rings. See Table IV for a description of the annealed case. Core and case hardness values and perhaps a more accurate method of obtaining hardened case depth are shown on the hardness curves in Figs. 3 to 8.

**Table VI**  
**Transverse Loading Test, Maximum Fibre Stress, and Deflection, as shown in Fig. I.**

Steel No.	Treat-ment No.	Stress at First Crack of Case Lbs. per Sq. In.	Deflec-tion at First Crack	Max. Load Lbs. per Sq. In.	Fracture
2	1	242,400	0.113"	242,400	Case and core rather coarse and crystalline.
2	2	304,000	0.150"	304,000	All fine and crystalline.
3	1	248,600	0.123"	248,600	Case rather coarse. Core partly silky.
4	1	289,800	0.147"	339,000	Fine and silky.
5	1	222,000	0.175"	222,000	Core mostly silky.
6	1	214,200	0.317"	214,200	Core coarse and crystalline.
6A	1	187,800	0.098"	255,000	Core coarse and mostly crystalline.
6A	2	No crack at	0.204"	247,000	Core coarse and crystalline.
7	1	212,000	0.147"	230,000	Fine and silky.
7	2	211,000	0.120"	260,500	Core fibrous or woody.
7	C-1	174,600	0.102"	187,100	Core partly crystalline.
7	C-2	198,000	0.127"	208,800	Core mostly silky.
7	C-3	219,500	0.121"	239,400	Core silky.
7	C-4	214,100	0.122"	262,600	Core fibrous or woody.
7	C-5	193,900	0.102"	201,000	Core partly crystalline.
7A	1	255,500	0.158"	266,000	Core silky.
7D	1	246,400	0.148"	265,000	Core silky.
9	1	264,600	0.175"	264,600	Fine and crystalline.
9	2	200,000	0.111"	248,900	Core silky.
9	3	277,700	0.140"	283,800	Core silky.
10	1	281,700	0.154"	286,500	Core silky.
10	2	196,400	0.190"	227,300	Core silky.
10	3	197,900	0.185"	232,000	Core silky.

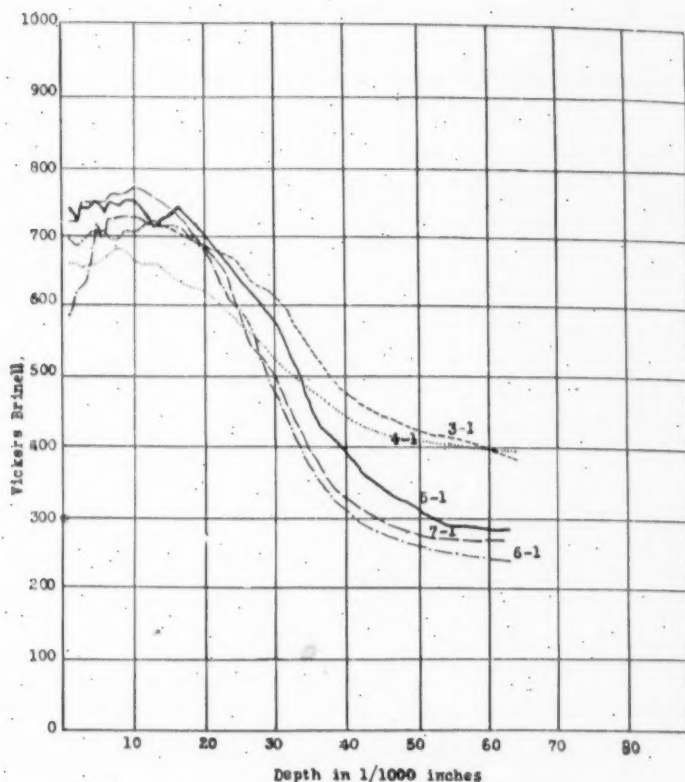


Fig. 3—Diagram Showing the Depth-Hardness Curves of Different Steels Carburized and Heat Treated Under Similar Conditions.

*Physical Tests.* The Izod bars were standard 3-notch bars 10 millimeters square and with 45-degree notches except that only two notches were made thus leaving space for one unnotched test on the same bars. The bars were machined, ground to size, and notched before carburizing. Duplicates were run on each test.

In the transverse load tests on the bars, deflection readings were taken at increments of 200 pounds. The results are given in Table VI. In most cases the first crack occurred at the elastic limit as closely as could be measured. Among bars of the same kind of steel given identical treatments, it was noticed that the load at which the first crack occurred was not necessarily an indication of the breaking load of that bar since frequently a bar would show a breaking strength equal to or greater than one in which the first crack of the case occurred at a higher load. This perhaps would not be the case under impact loads and quite probably would not under fatigue failure.

Table VII  
Carburized Izod Test Results

Steel No.	Treatment No.	Notched 1.5 ft. lbs.	Unnotched 7.5 ft. lbs.	Fracture
2	1			Coarse case. Crystalline core. Cooled in still air from 1450°. Case and core fine and crystalline.
2	2	1.5 ft. lbs.	15 ft. lbs.	Crystalline case. Silky core.
3	1	6.5 ft. lbs.	14 ft. lbs.	Fine case. Silky core.
4	1	12 ft. lbs.	30 ft. lbs.	Fine case. Fine crystalline core.
5	1	5 ft. lbs.	11.5 ft. lbs.	Fine case. Crystalline core.
6	1	9.5 ft. lbs.	13 ft. lbs.	Fine case. Silky core.
7	1	23 ft. lbs.	47 ft. lbs.	Fine case. Crystalline core.
7	C-1	5 ft. lbs.	11 ft. lbs.	Fine case. Crystalline core.
7	C-2	7 ft. lbs.	11.5 ft. lbs.	Fine case. Core mostly silky.
7	C-3	16 ft. lbs.	19.5 ft. lbs.	Fine case. Core silky.
7	C-4	18 ft. lbs.	26 ft. lbs.	Fine case. Crystalline core.
7	C-5	5 ft. lbs.	10.5 ft. lbs.	Fine case. Core silky.
7A	1	6 ft. lbs.	17 ft. lbs.	Fine case. Case and core fine and crystalline.
7D	1	3 ft. lbs.	12.5 ft. lbs.	Cooled in still air from 1400°. Case and core fine and crystalline.
9	1	2 ft. lbs.	17 ft. lbs.	Fine case. Silky core.
9	2	5.5 ft. lbs.	16 ft. lbs.	Fine case. Core partly silky.
9	3	5 ft. lbs.	18 ft. lbs.	Fine case. Silky core.
10	1	7.5 ft. lbs.	28.5 ft. lbs.	
10	2	15 ft. lbs.	25 ft. lbs.	
10	3	24 ft. lbs.	40 ft. lbs.	

Stress-strain readings were taken on the rings loaded as in Fig. 2. Readings were taken at increments of 500 pounds and the results given in Table VIII. The load was applied only until the case cracked. The unbroken halves were then tested for impact resistance. These parts, of course, were not standard specimens, but since they were all the same size, the results give comparative unnotched transverse impact values on the case hardened pieces. The rings in all cases were made from cross sections of bar stock and were not given any forging operations. These impact results are given in Table VIII.

The depth-hardness curves shown in Figs. 3 to 8 are averages of two sets of readings taken on parts of the impact plugs after grinding to a 4-degree taper. The first ten readings near the outside of the case were taken 1/60-inch apart horizontally, or approximately 0.001-inch increase in depth, each of the next ten at 0.0035-inch increase in depth and the remainder at 0.007-inch increase. The hardness readings were obtained by the Vickers diamond indenter. For a comparison of Vickers hardness with Rockwell, scleroscope and Brinell hardness, see a previous article by one of the authors.<sup>1</sup>

*Microscopic Examinations on Treated Test Bars.* Steel 3 and 3A (S.A.E. 2315). Case martensitic and shows no excess carbide.

<sup>1</sup>O. W. McMullan, "Effect of Quenching Temperature Change on the Properties of Quenched Steel," TRANSACTIONS, American Society for Steel Treating, Vol. XIV, October, 1928, p. 477.

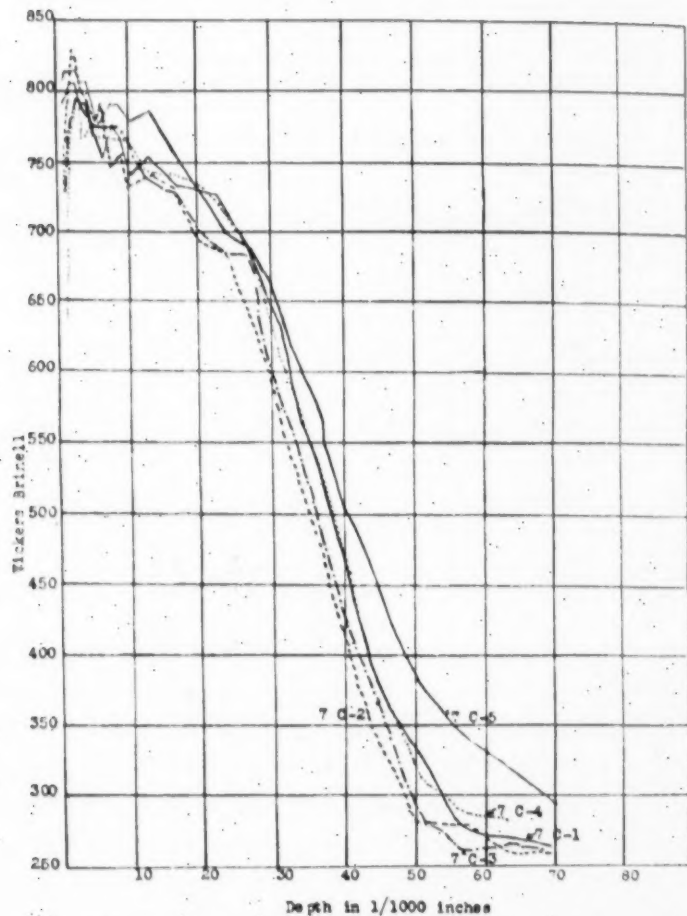


Fig. 4—Diagram Showing Depth-Hardness Curves and Effect of Different Times and Temperatures of Cyanide Bath Heating. Low Core Hardness is Due to Low Carbon Content of Steel.

Core in the quenched condition. Case and core rather coarse-grained.

Steel 4 and 4A (S.A.E. 2512). Martensitic case with no excess carbide. Quenched core. Fairly fine to average grain size.

Steel 5 (S.A.E. 3115). Martensitic case with a little excess carbide. Core not completely quenched showing some ferrite. Fine case, fairly fine core.

Steel 6 (Special 6115). Izod bar and rings show a martensitic case with a little excess carbide and some ferrite in the core. Fine case and core. The transverse bar was troostitic in the outer part of the case with a narrow martensitic band underneath and had considerable ferrite in the core.

Steel 6A (Special 6115). The samples quenched from the



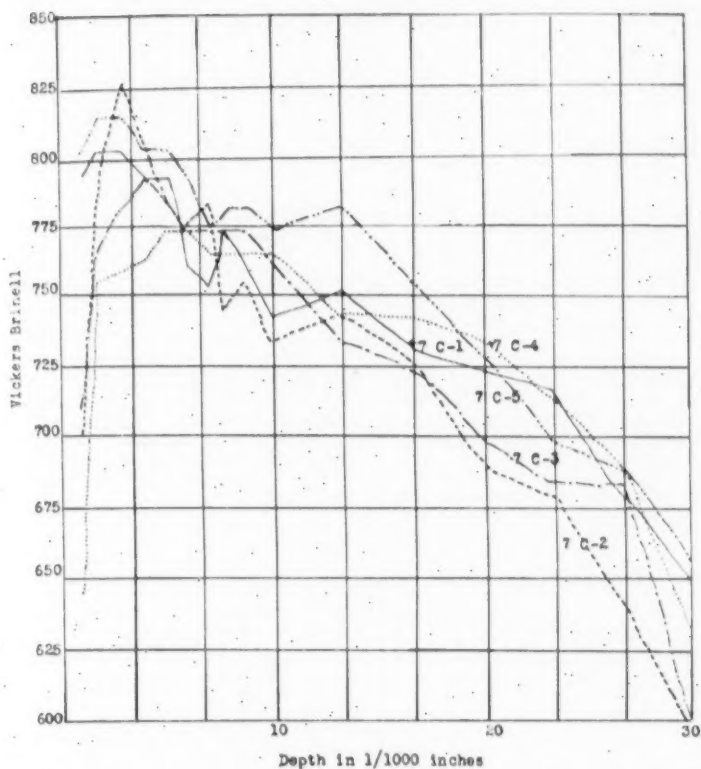


Fig. 5—Same as Fig. 4 but on Enlarged Hardness Scale.

carburizing pot show a rather coarse-grained acicular martensitic case and a quenched core. The sample quenched from 1450 degrees Fahr. (790 degrees Cent.) was troostitic in the outer part of the case and martensitic in the eutectoid portion. The core contained much free ferrite.

Steel 7 (S.A.E. 4615). After 1525 degrees Fahr. (830 degrees Cent.) oil quench, the case was martensitic with no excess carbide. Very little ferrite in the core. Fine-grained case and core. After quenching from the pot, the structure was similar except there was no free ferrite in the core.

All the cyanide-treated samples have a fine-grained martensitic case, those quenched from 1410 degrees Fahr. (765 degrees Cent.) showing excess carbide and core containing free ferrite. Those quenched from 1510 degrees Fahr. (820 degrees Cent.) were martensitic in the case with no excess carbide. The core of the five minute samples contained free ferrite while those heated 30 minutes were completely quenched.

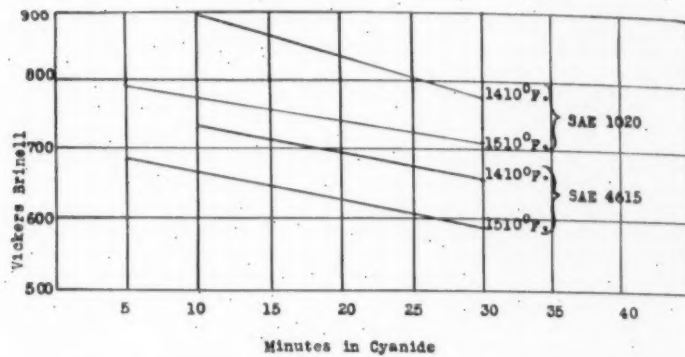


Fig. 6—Diagram Showing Surface Hardness of Carburized S. A. E. 4615 and Carburized S. A. E. 1020 Steel Heated at Different Temperatures and Times in Cyanide Pot. Hardness of Steel 7 Quenched in Water After 10 Minutes in Cyanide at 1410 Degrees Fahr.—765. Steel 7A Quenched into Oil from Furnace at 1525 Degrees Fahr.—760. Steel 7D Quenched into Oil from Furnace at 1525 Degrees Fahr.—723.

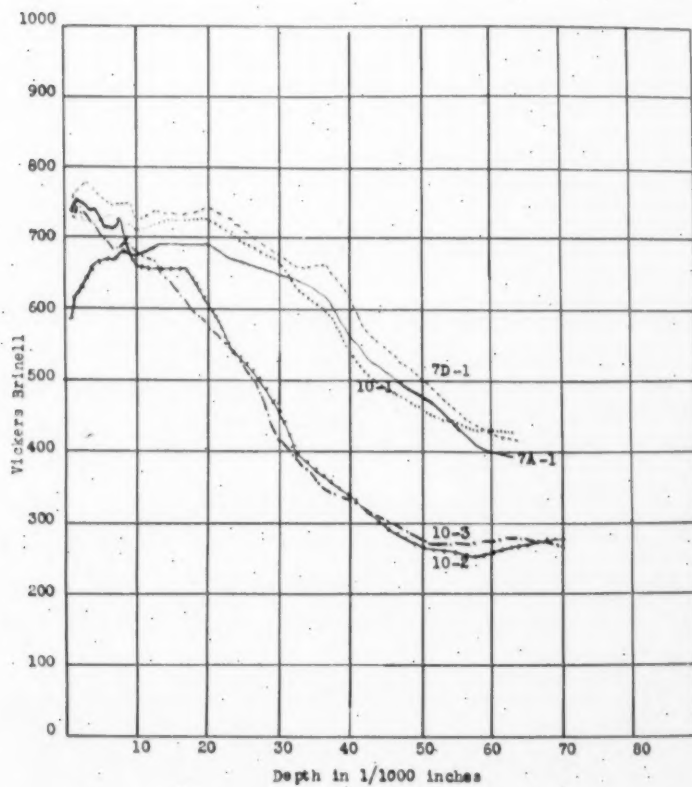


Fig. 7—Diagram Showing Depth-Hardness Curves of Case Hardened S. A. E. 4615 and Case Hardened S. A. E. 4615 Densite Treated Steel in the Ladle. Also Oil-quenched and Air-cooled Carburized 3.50 per cent Nickel-molybdenum Steel. Note Higher Core Hardness of S. A. E. 4615 as Compared with Figs. 3 and 4. This is due to Higher Carbon Content in Core.

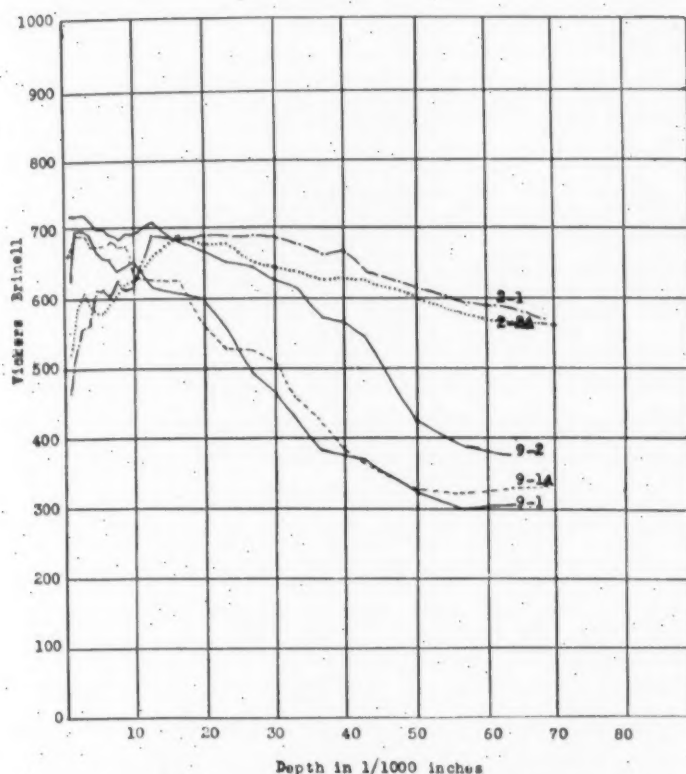


Fig. 8—Diagram Showing Depth-Hardness Curves of Carburized and Air-cooled 0.40 Per Cent Carbon-chromium-nickel-molybdenum Steel, and Carburized and Treated High Manganese-molybdenum Air-cooled and Oil-quenched. The Numbers on the Curves of Figs. 3 to 8 Inclusive Refer to the Steel and the Treatment Number Given in Tables I, II, and III.

Steel 7D (Densite treated S.A.E. 4615). Similar to the same treatment on S.A.E. 4615.

Steel 9 (High manganese-molybdenum). Cooled in air from pot. Case mostly martensitic with a little troostite and a fine quenched core. Those cooled in air from a lower temperature had a similar structure.

The oil-quenched samples showed a martensitic case and quenched core.

Steel 10 (High nickel-molybdenum). Cooled in air from pot. The case shows a martensitic structure mixed with some troostite. The outer part shows a fine acicular martensitic structure in a ground mass having the appearance of austenite, which remains light on repeated etchings. Core very fine, but contains free ferrite. Those cooled in air from a lower temperature show a mostly martensitic case with some troostite and a core similar to the above. The oil-

quenched samples show a martensitic or martensitic-troostitic case and a quenched core.

Steel 2 (Chromium-nickel-molybdenum). The samples air-cooled from the pot show coarse acicular martensite in an austenitic appearing background in the outer part of the case, the background remaining white after repeated etchings. This structure is more pronounced than in steel 10, and was shown to a small extent by one sample of steel 9. The presence of austenite is shown by the hardness curves in Figs. 7 and 8. The hardness near the surface is lower than the inner part of the case. The core was coarse and martensitic troostitic. The samples cooled in still air from 1450 degrees Fahr. (790 degrees Cent.) were fine-grained, showed less of the austenitic structure, and a martensitic core. The sample given an air blast from 1450 degrees Fahr. was fine-grained martensite throughout, without any austenitic appearing structure in the case.

Steels 6 and 6A are special 6115, that is, chromium-vanadium containing about one-half per cent of nickel and chromium. Steel 6 was treated first and as shown in Tables IV and V the hardness of the case was unsatisfactory. A different heat of the same type of steel (6A) was then used and similar results obtained. Samples of this steel were quenched from 1600 degrees Fahr. (870 degrees Cent.) into oil, both from an electric tool hardening furnace and an oil-fired production hardening furnace. Samples were also quenched in oil from an electric furnace at 1400 to 1475 degrees Fahr. (760 to 800 degrees Cent.) In none of these attempts were successful results obtained as far as satisfactory surface hardness was concerned. Rockwell readings taken on the polished quenched pieces showed many places as low as 39 to 42 Rockwell. Samples of this analysis were quenched from a cyanide pot at 1500 degrees Fahr. (815 degrees Cent.) in oil and water and some were also quenched directly from the carburizing pot into oil. Those samples quenched in water or from the carburizing pot into oil were the only ones giving satisfactory hardness. The water quench would, of course, be eliminated for a gear treatment. The hard samples showed a martensitic case throughout, while the others were troostitic on the outside and martensitic in the eutectoid zone. The samples contained excess carbide in the outer part of the case as did the other steels, and were not unusually spheroidized for a vanadium steel. No explanation has been found for the poor results.



1929

Table VIII

Physical Properties of Test Rings When Loaded on a Point on the Circumference as Shown in Fig. 2. The Impact Results Were Obtained on the Same Rings and are Merely Comparative for Transverse Impact Values

Steel No.	Treatment No.	Load at First Crack of Case	Deflection at First Crack 0.015"	Unnotched Impact 13.5 ft. lbs.	Fracture
2	1	9,500 lbs.	0.015"	13.5 ft. lbs.	Case rather coarse. Crystal-line core. Not banded.
2	2A	12,500 lbs.	0.025"	19 ft. lbs.	Cool in still air from 1450°. Fine and crystalline.
3A	1	10,500 lbs.	0.0165"	17 ft. lbs.	Case rather coarse. Core silky, fibrous and banded.
4A	1	14,000 lbs.	0.024"	40 ft. lbs.	Core silky and banded.
5	1	9,000 lbs.	0.015"	14 ft. lbs.	Mostly crystalline. Not much banding.
6	1	11,700 lbs.	0.022"	21.5 ft. lbs.	Mostly crystalline. Not much banding.
7B	1	10,500 lbs.	0.017"	22 ft. lbs.	Core silky and banded.

Table IX

Dimensional Change After Carburizing and Quenching

Steel No.	Treatment No.	Change in Diameter from Original		Change in Length from Original	
		After Carb.	After Treat.	After Carb.	After Treat.
2	1	.....	+ .00125"	.....	-.0020"
2	2*	.....	+ .00082"	.....	.....
3	1	+ .0008"	+ .00055"	-.0014"	+ .0008"
4	1	+ .00087"	+ .0006"	-.00075"	+ .0061"
5	1	+ .00083"	+ .00048"	-.00233"	+ .00013"
6	1	+ .00075"	+ .0001"	-.0019"	+ .00213"
6A	1	+ .00096"	.....	-.00483"	.....
6A	2	+ .00108"	-.0038"	-.0060"	-.0050"
7	1	+ .00093"	+ .00036"	-.00243"	-.0026"
7	2	+ .00097"	+ .00097"	-.00533"	-.0048"
7	C-1	+ .00093"	+ .00128"	-.00327"	-.00533"
7	C-2	+ .00082"	+ .00105"	-.00167"	-.0045"
7	C-3	+ .00095"	+ .0014"	-.0015"	-.00943"
7	C-4	+ .00105"	+ .00167"	-.0020"	-.01433"
7	C-5	+ .00092"	+ .00135"	-.00233"	+ .00107"
7A	1	+ .00062"	+ .0009"	-.00233"	-.01183"
7D	1	+ .00085"	+ .00065"	-.00133"	-.0061"
9	1	.....	+ .00072"	.....	-.00683"
9	2	+ .00085"	-.00056"	-.00415"	-.00548"
9	3***	.....	-.00237"	.....	-.00130"
10	1	+ .00108"	+ .0010"	-.0055"	-.0091"
10	2	.....	+ .00015"	.....	+ .00075"
10	3****	.....	-.00005"	.....	-.0080"

\*After several reheatings.

\*\*After several treatments and recarburizing.

\*\*\*Reheated to 1450 degrees Fahr. and oil-quenched after treatment 2.

\*\*\*\*Reheated to 1450 degrees Fahr. and air-cooled after treatment 2.

It had been found in practice that cyaniding carburized S.A.E. 4615 steel produced a soft skin on the treated part which could be filed off and that this layer became deeper the longer the time in cyanide. These results were confirmed at both the high and low temperatures, the effect apparently being somewhat greater at the high temperature. The intention was to run the short-time samples

**Table X**  
**Change in Dimensions of Rings After Carburizing and Treating**

Steel No.	Treat-ment No.	Change in Outside Diameter from Original		Change in Inside Diameter from Original		Change in Thickness from Original		Eccentricity	
		After Carb.	After Treat.	After Carb.	After Treat.	After Carb.	After Treat.	After Carb.	After Treat.
2	1	.....	+ .00043"	.....	-.0006"	.....	+ .00103"	.....	.00015"
3	1	.....	-.00032"	.....	-.00318"	.....	-.00615"	.....	.00105"
4	1	.....	+ .00005"	.....	-.0017"	.....	+ .00034"	.00055"	.0012"
5	1	.....	-.00055"	.....	-.0018"	.....	+ .00003"	.0004"	.0008"
6	1	.....	-.00043"	.....	-.00375"	.....	-.00032"	.00055"	.00095"
7	1	.....	+ .00103"	.....	-.0014"	.....	-.00025"	.00015"	.0009"
		.....	-.00138"	.....	-.0018"	.....	-.00037"	.0004"	.0009"
		.....	+ .00065"	.....	-.0053"	.....	.....	.....	.....

**Table XI**  
**Comparative Core Properties of Steels 7D, 9 and 10**

Steel No.	Elastic Limit Lbs./Sq. In.	Maximum Lbs./Sq. In.	% Elong.	% R.A.	Plug	Brinell Surface of 1" Round	Izod	
							Ft. Lbs.	Brin.
7D	47,500	123,600	25.4	55.6	262	262	52	241
9	90,000	176,400	16.8	62.8	351	364	45	364
10	45,000	128,750	22.6	66.5	273	332	85	258

All samples treated in 1-inch round sections normalized at 1650 degrees Fahr. and oil-quenched from 1500 degrees Fahr. No temper given.

at 1410 degrees Fahr. (765 degrees Cent.) for five minutes, but it was found necessary to give them about ten minutes to come up to heat. The results indicate that unless very carefully controlled that cyaniding is a disadvantage rather than an advantage to the surface hardness obtained on such carburized work. As has previously been found, the lower quenching temperatures gave higher hardness values. The depth hardness curves on the cyanided samples are shown in Figs. 4 and 5, and surface hardness readings in Fig. 6. Surface hardness results on carburized S.A.E. 1020 bars 1 inch in diameter heated in cyanide and water-quenched are also shown in Fig. 6. The Rockwell hardness shows some difference, but the impressions are too deep to show the effect on the surface.

Apparently the variation in case depth did not have much effect on the transverse strength of the bars. A more important factor in obtaining good transverse strength seems to be the heat treatment given the bars. The breaking load is also greatly affected by the structural condition of the core and it seems to be important that a uniformly well quenched core be obtained if possible. The Brinell hardness, while it usually indicates the relative core strength, is not a satisfactory index of the resistance to cracking of a carburized part. Increase in carbon content of the core apparently increased the

load-carrying capacity. Difference in type of steel and the treatment given the same type of steel were much more pronounced in effect on the impact values than was the core hardness. Izod bars showing high core hardness frequently gave much better impact values than those of low hardness of a different analysis or treatment. It is quite probable that case depth, grain size and hardness of the case have considerable effect on the shock resistance. Different steels and treatments seem to give different relative values between notched and unnotched impact tests.

The dimension changes given in Tables IX and X present some interesting and valuable information from the standpoint of gear cutting. It can be seen that some steels show an expansion where others show a contraction. In some cases hardening intensifies the change started by carburizing and in others reverses it; and different treatments of the same steel produce widely differing results. Since in spiral gear tooth cutting, allowance is frequently made for bearing changes after carburizing and hardening, satisfactory bearing locations cannot be expected if a change is made in the type of steel used or even of the heat treatment given the same steel, unless the differing dimension changes and warpage are taken into consideration. The test bars used were not entirely satisfactory for warpage test, but indicate that the air-cooled samples offer more possibilities in the way of simplicity of treatment rather than elimination of warpage. There is need for further investigation of some of these types of steel.

#### SUMMARY

##### *Physical Properties*

The results indicate that of the steels tested, the S.A.E. 4615 will give the hardest surface when oil-quenched. This would indicate that this steel is well suited to resist wear in the case hardened condition. The addition of nickel tends to lower the surface hardness obtained by oil quenching, and the 5 per cent nickel steels are low in this respect. The impression which is prevalent that the surface hardness can be increased by heating carburized parts in a cyanide bath does not check up and the results indicate that quite the reverse is true. The heating in cyanide must be carefully regulated so that the work is in the heating bath a minimum of time, inasmuch as there is a decided decrease in surface hardness as the time or temperature in the cyanide bath is increased.

In connection with the hardness of steels, it is believed that the depth-hardness curve as obtained by a Vickers diamond on tapered sections is extremely valuable and warrants a more extended use than at present. It has been found to give information which it is very difficult to obtain by any other method.

Steels such as the high manganese-molybdenum, etc., which in the carburized condition have the  $A_{r1}$  point suppressed so that martensite is formed on air cooling, show good hardness readings and it is probable that these steels may find a very important application. The characteristics of the high manganese-molybdenum steel No. 9 after carburizing and air cooling are very satisfactory. The core hardness is not too high and the strength of the case compares favorably with the 3.50 per cent nickel steels, and the physical properties of the core are much better than the 3.50 per cent nickel steels. The 3.50 per cent nickel-molybdenum steels do not compare well with the standard carburizing steels in the air-cooled condition, although in the oil-quenched condition they have properties which compare well with the 5 per cent nickel steels, and it is possible that this steel may be developed to compete with the 5 per cent nickel steels for gear service.

The S.A.E. 2512 steel is apparently the outstanding steel for load carrying capacity under static conditions. Apparently in any one type of steel those treatments tending to give a homogeneous core structure, give the highest case strength figures. The structure of the core is apparently of more importance than the hardness of the core. For this reason, the higher carbon cores, possibly because of great ease in obtaining uniform core structure, give higher load readings.

The resistance to notch propagation as indicated by the Izod values for steels quenched above the critical range of the core, has the highest values in the nickel-molybdenum steels, particularly the S.A.E. 4615. This checks with results in service where it has been found that under extremely severe conditions such as in super-charger gears, the S.A.E. 4615 gives most satisfactory results. As shown by the cyanide treated samples quenched from different temperatures, core structure is more important than core hardness as far as the Izod values are concerned.

Considerable variation in dimension changes was found to exist among different types of steel and different treatments of the same type. Hardening in some cases intensifies the carburizing changes



and in others reverses it. In spiral gear tooth cutting where allowance is made for change in location of bearing after carburizing and hardening these changes are important to take into consideration where a change in type of steel or heat treatment is to be made. The ring tests showed a slight decrease in both inside and outside diameter after treatment, with one exception on the outside diameter. Apparently the air-cooled steels would be the most satisfactory for minimum change in bearing.

In brief it may be said that where high surface hardness and high resistance to impact are desired, together with good strength; S.A.E. 4615 single-quenched from above the upper critical range of the core gives the most satisfactory results. Where extremely high load carrying capacity and good impact resistance are desired and extreme surface hardness is not necessary, S.A.E. 2512 offers the best combination of properties.

High alloy air-hardened steels offer interesting possibilities, chiefly in regard to simplicity of treatment. The results here given are merely preliminary as there was no background of commercial experience.

The above results do not include many important items in gear manufacture. The interesting results obtained with the manganese and manganese-molybdenum steels have also not been included. The preliminary results obtained with these steels indicate that they will take their place in ring gear and pinion manufacture in direct competition with S.A.E. 2315 and S.A.E. 4615. This investigation has developed the great need of further investigation into phases of the steels, treatments, etc., not at present well developed. There is great need for an investigation into the relation between the grain size, composition, etc., and the machining results. There is also great need for an investigation into the effect of carbon content of the core on the physical properties of the finished gear, and very valuable papers could be written on the manganese and manganese-molybdenum steels. It is the hope of the writers that such papers will be published so that the gear industry will have available this much needed information.

#### DISCUSSION

**Written Discussion:** By Jerome Strauss, Vanadium Corporation of America, Bridgeville, Pa.

The authors have presented an excellent study of those properties

of case hardening gear steels determinable in the laboratory. It is unfortunate, however, that some of the steels which they mentioned or discussed briefly, but did not report upon, have not been included in these tests in order that this single paper would then be a complete presentation.

They have referred to the possible utility of manganese steels for these applications. Simple pearlitic manganese steels with manganese above 1 per cent have not found use as gears and parts subjected to similar stresses due to low impact strength with low tempering temperatures and high hardnesses. The addition to these steels of another element, suitably selected, improves the impact strength, but unfortunately the authors have dealt with only one of these and then not only briefly, but with a composition that may prove to be dangerously high in manganese. Reference has also been made in the paper to S.A.E. 6115 as one of the three principal steels used for case hardening gears. But, unfortunately, this steel has not been included in this report. It is also noticeable that no intermediate or high chromium-nickel steels have been included. One vanadium steel which has been tested has been called "Special 6115". Such a term implies a material representing an improvement over the standard type. Actually, however, the steel appears to be not nearly equal to the standard type, at least in such sections as are represented by the authors' test specimens. The authors state "No explanation has been found for the poor results," but when a steel is used of such composition that it is just able to yield satisfactory results on very small sections such as passenger car gears (and this it does consistently) and then is compared with other steels—all being tested in  $\frac{1}{2}$ -inch sections—there is surely a full explanation. In other words, for the sections of the test pieces employed, the material in question is a water hardening steel and it is hopeless to try to secure comparable properties by oil hardening. This is effectively brought out by a study of the core and case hardnesses of Table IV, especially test 6 and test 6A—treatment 2. It will be observed that while the case hardnesses are low, the core hardnesses are also extremely low—in fact, lower than in any of the other steels. Also the easy cracking on small deflection of test 6A treatment 1 is not surprising for such a treatment; in this connection, note the results with the steel which the authors recommend as best for this service (No. 7) after having received like treatment.

Some irregularity in the results of these particular tests may be expected. No mention was made in the paper of whether the test pieces were merely turned or whether they are ground, or ground and polished. In tests of this character, made on hardened and hardened and tempered steels, the writer has observed large differences between duplicate tests unless polishing or lapping was resorted to.

It is interesting to note the authors' comments to the effect that cyaniding yields a soft skin on S.A.E. 4615. Long experience with S.A.E. 6115 and the "Special 6115" of this paper has not shown such results. Of course, it is stated in the paper that there was used "96 per cent cyanide except for decomposition through use." No mention is made

of the degree of decomposition but, as either carburizing or decarburizing may occur in atmospheric furnaces according to the temperature and concentration of each gas present and also the steel composition, is it not possible that similar occurrences may be observed in the cyaniding bath?

The writer is also very much interested in the comments on the variations in open-hearth and electric steels. While many different features of practice no doubt have some influence on the properties of the finished product, it may well be asked whether consideration has been given to the size of the heat as influencing the number of ingots, and hence the pouring temperature variation, and therefore the solidification time of the ingot with all the known effects of the latter upon the properties of steels. In general, electric furnaces are small, although there are now installations of sufficient size to make possible an approach, at least, to the answer to this question. From another view point, while much progress has been made by the steel maker, there appears still to be a great deal more to be accomplished in the control of forging temperature heating rate and forging rate. When this is done in the average drop forging plant, troubles resulting from nonuniformity will probably suffer a great decrease.

#### Oral Discussion

G. M. EATON: I would like to ask a question in connection with the Izod test. The physical ratio of the face width of the gear tooth to the thickness of the tooth is radically different from the ratio between the width and thickness of the Izod test piece. Even with a material which is quite uniform throughout its cross section, the width of the Izod test piece is important, particularly when the fracture shows diagonal shear at the edges. My question is this: "Does the nitrided notched Izod test piece show a square break without the ears that occur at the sides of ductile specimens"?

O. W. McMULLAN: The break is a straight-across break. I imagine that was rather to be expected because of the high hardness of the case.

G. M. EATON: It would seem to me that in carrying a study of this sort further it would be an entirely justifiable development to make the Izod test piece about 3 times as wide and not compare it with results on standard Izod test pieces. This will minimize the effect of the support which the case offers as compared with the object that is being tested. The purpose of this of course is to secure a better comparison of various nitrided steels and nitriding cycles as applied to gearing.

O. W. McMULLAN: I believe that is a good point; however, on the gear teeth that Mr. Eaton has mentioned, it is usual to have the ends of the teeth carburized so that we have some support at least.

G. M. EATON: But the ratio of case to core plus core is different in the Izod piece from the ratio existing in the gear.

O. W. McMULLAN: Yes, it is. Of course, the carburizing of these steels decreases the impact greatly, and a greater percentage of core probably gives you higher impact results rather than lower so I do not know whether the support of the case is a very important factor in that way or not.

H. B. KNOWLTON: I would like to ask the author a few questions with regard to the correlation of his test with surface failures of gears. Which of these tests or combinations of tests gives the best indication of how a gear will stand up in service? What is the usual cause of spalling in the case of the gear? Is it due to the case being too thin to stand the load, consequently being dented in and spalling, or is it due to inherent brittleness of the case due to too coarse a grain structure or excess cementite in an envelope form? When a tooth breaks out by the root is the usual explanation of this due to a fatigue failure starting in the case? May it be attributed to the structure of the case, or is it due to inherent weakness of the core?

H. K. HERSCHMAN: I should like to comment on two points in this paper. The first has to do with quenching the specimens from the carburizing box. Some years ago the speaker worked on the investigation of carburizing armor plate and it was found that the "box-cooled" plate when quenched from a temperature even as high as the carburizing temperature, showed considerable cementite. That is, the quenched specimen examined under the microscope still showed the existence of the cementite envelopes commonly seen in plate which has been carburized and slowly cooled. In specimens which had been similarly carburized and subsequently rolled at 1100 degrees Cent. (2010 degrees Fahr.) and then quenched, the existence of the envelopes was not observed. This raised the question whether the specimens carburized should not be normalized at a temperature somewhat in excess of the carburizing temperature to completely dissolve the precipitated cementite, a great deal of which appears as envelopes. Of course, in the experiments which were carried out at the Bureau and to which I have reference, the envelopes were perhaps much more noticeable than in the case of the specimens of which Mr. McMullan spoke because in the former the carburizing periods were as high as 72 hours which developed very large and readily noticeable envelopes, observable even at very low magnifications. This may also have some reference to Mr. Knowlton's question as to the spalling often experienced in carburized steels. We have found in testing armor plates that spalling often did occur when cementitic envelopes were present. However, the specimens which were heated above the carburizing temperature, normalized and then quenched showed little or no precipitated cementite and spalling was not experienced. I should also like to make reference to Mr. McMullan's impact results. I gathered from his statements that these tests were made on specimens carburized and not subsequently heat treated. In his deeply carburized specimens the impact values were very low as compared with those having a lesser depth of carburization. I was wondering whether that did not also tie up with the question of the existence of envelopes in the deeper carburized specimens. If this wide difference was partly due to spalling in the deeper carburized specimens, then from the discussion above, it would appear that proper treatment for complete solutions of the carbides would produce results which would be less divergent than those obtained by Mr. McMullan.

JEROME STRAUSS: I would like to refer again to the question of section.



Mr. McMullan was kind enough to let us have some of the specimens of the "Special 6115" which had lower chromium than the standard composition and some nickel. He mentioned that he was able to secure full hardening when quenching from the pot, but not on quenching after reheating to either 1450 or 1600 degrees Fahr. It is probably a question not only of section but also of time at the quenching temperature and the effecting of full solution of carbide as influenced by cooling rate after carburizing and combined with atmosphere conditions in heating for quenching. We took the specimens that he provided and after turning off the case, recarburizing at 1725 degrees Fahr., pot cooling and reheating to 1525 degrees Fahr. they hardened fully in oil after 10 minutes at temperature.

O. W. McMULLAN: Mr. Knowlton has asked some questions which have been the subject of argument for a good many years and I am hardly prepared to stand up here and give any definite information on these questions.

With regard to the service of the gear, of course the hardness is one thing which is very essential to the wear resistance and of course there are other considerations to be taken in, as foreign material and oil and such items as that. As far as spalling, I do not think I can add anything to that subject, probably one of the things is the coarse grain structure, and also the presence of envelopes of excess carbide. In the breaking out of the tooth at the root, I mentioned one thing somewhat earlier which I believe has an affect on that. It is the method of forging a ring gear. There are two or three common methods; one is to simply forge the gear from flat stock, another is the split ring, and the method we use is upsetting a bar four or five inches square into a pancake and forging the gear from this. In the last case we have forged flow lines in such a position that it greatly increases the strength, particularly of the core while this condition does not prevail in gears forged from flat stock. We have been troubled with practically no breaking out of gear teeth at the root by forging gears in this method. We find that the radius at the root of the tooth is an important factor in starting fatigue failures and we are particularly careful to obtain as large a radius as possible. As far as case depth is concerned, the question of type of service I think would be important. Where there is a shock load the thinner case would give higher impact value, while with the steady load the higher stress would be obtained by the deeper case.

In regard to the impacts on the carburized specimens those which I gave results all were heat treated before testing for 8 and 16 hours carburizing and quenched directly from the carburizing box, were S.A.E. 4615 and the microsections show no excess carbide. The same was also true of the S.A.E. 6115 steel quenched direct, that is the special 6115 steel quenched direct from the carburizing box. None of the samples quenched from the box showed excess carbide, that being present only in some of the samples quenched at low temperatures. I might say a word more about the special S.A.E. 6115 in answer to Mr. Strauss in regard to the length of time in which these samples were heated. They were first heated in an electric furnace at 1600 degrees Fahr. and they were held in there a long enough time at that temperature so that they were badly scaled and I thought at first the low hardness was due

**Table Ia**  
**Analysis of Steels**

Steel No.	C	Mn	P	S	Mo	Si	Grain Size	Source
1A	0.21	0.41	0.023	0.035	.....	0.18	9	Open-Hearth
1B	0.21	0.50	0.015	0.040	.....	0.17	1-3	Open-Hearth
11	0.16	1.62	0.018	0.023	.....	0.12	7	Electric Furnace
12	0.15	2.00	0.020	0.021	0.38	0.10	7	Electric Furnace

**Table IIa**  
**Heat Treatment of Samples Tested**

Steel No.	Treatment No.	Normalized Deg. Fahr.	Carburized Deg. Fahr.	Cool from Carburizing	Quenched Deg. Fahr.	Tempered Deg. Fahr.
1A	1	1650	1675 12 hrs.	Pot	1430 Water	375 20 min.
1A	2	1650	1675 12 hrs.	Pot	1575 Wat.-1430 Wat.	375 20 min.
1B	1	1650	1675 12 hrs.	Pot	1430 Water	375 20 min.
1B	2	1650	1675 12 hrs.	Pot	1575 Wat.-1430 Wat.	375 20 min.
7	3	1650	1675 8 hrs.	Pot	1525 Oil	300 15 min.
7	4	1650	1675 16 hrs.	Pot	1525 Oil	300 15 min.
7	2	1650	1675 12 hrs.	Oil		300 15 min.
7A	2	1650	1675 12 hrs.	Oil		300 15 min.
11	1	1650	1675 12 hrs.	Pot	1525 Oil	300 15 min.
12	1	1650	1675 12 hrs.	Pot	1525 Oil	300 15 min.

**Table IVa**  
**Showing Case Depth of Pot Cooled and Treated Samples and Surface Hardness of Treated Bars for Transverse Test**

Steel No.	Treatment No.	Hyper	Case Depth on Pot Cooled Samples Saturated	Total	Hardened Case Depth on Test Bar	Rockwell on Case	Hardness Brinell on Core
1A	1	0.015"	0.028"	0.056"	0.043"	62.5-65	156
1A	2	0.015"	0.028"	0.056"	0.041"	59 -64	156
1B	1	0.018"	0.034"	0.065"	0.043"	62.5-65	156
1B	2	0.018"	0.034"	0.065"	0.043"	60 -64	156
7	3	0.016"	0.027"	0.045"	0.037"	61 -62.5	223
7	4	0.027"	0.045"	0.072"	0.065"	60.5-62	235
7	2	0.021"	0.036"	0.059"	.....	.....	.....
7A	2	0.027"	0.038"	0.067"	0.055"	57 -62	255
11	1	0.026"	0.048"	0.070"	0.045"	53 -61	207
12	1	0.021"	0.047"	0.070"	0.060"	58 -61	262

**Table VIa**  
**Transverse Loading Test, Maximum Fibre Stress, and Deflection, as Shown in Fig. 1**

Steel No.	Treatment No.	Stress at First Crack Lbs. Per Sq. In.	Deflection at First Crack	Max. Load Lbs. Per Sq. In.	Fracture
1A	1	191,900	0.125"	191,900	Case fine. Core fairly fine and crystalline.
1A	2	211,300	0.138"	211,300	Case and core fine and crystalline.
1B	1	146,450	0.076"	146,450	Case mostly fine. Core coarse and crystalline.
1B	2	189,500	0.130"	189,500	Case fine. Core fairly fine and crystalline.
7	3	209,700	0.160"	234,500	Case fine. Core silky fibrous and woody.
7	4	261,400	0.170"	264,800	Case fine. Core silky fibrous and woody.
7A	2	202,100	0.104"	264,000	Case fine. Core silky fibrous and woody.
11	1	211,900	0.177"	212,800	Case fine. Core crystalline.
12	1	225,000	0.126"	237,000	Case fine. Core mostly crystalline.

Table Va and VIIa  
Case Depth, Rockwell Hardness and Izod Values on Carburized Izod Bars

Steel No.	Treatment No.	Rockwell on Case	Hardened Case Depth	Notched Foot Pounds	Un-notched Foot Pounds	Fracture	
1A	1	64.5-66	0.048"	2	8	Case fine, coarse.	Core crystalline. Rather
1A	2	60.5-63.5	0.045"	6.5	8	Case fine.	Core fine and crystalline.
1B	1	64-65.5	0.046"	1	3	Case partly coarse.	Core coarse and crystalline.
1B	2	61-65	0.048"	5	7.5	Case and core fairly fine.	and crystalline.
7	3	62-63	0.055"	12.5	20	Case fine.	Core partly crystalline.
7	4	61.5-62	0.090"	4	8	Case fine.	Core partly crystalline.
7	2	60-62.5	0.055"	8	15.5	Case fairly fine.	Core silky fibrous.
7A	2	60-63	0.057"	6	13.5	Case fine.	Core silky.
11	1	60-61	0.060"	1.5	8	Case fairly fine.	Core crystalline.
12	1	60-62	0.056"	2	10	Case fine.	Core crystalline.

Table IXa  
Dimensional Change After Carburizing and Quenching

Steel No.	Treatment No.	Change in Diameter from Original		Change in Length from Original	
		After Carb.	After Treat.	After Carb.	After Treat.
1A	1	+ .00087"	+ .0012 "	+ .0015 "	- .0010"
1A	2	+ .00081"	- .00009"	+ .00133"	- .0150"
1B	1	+ .00107"	+ .0009 "	- .0030 "	- .0142"
1B	2	+ .00095"	- .00014"	- .0020 "	- .0120"
7	3	+ .00383"	+ .00015"	- .0018 "	- .0027"
7	4	+ .00129"	+ .00112"	- .0003 "	- .0073"
7A	2	.....	+ .00133"	.....	- .0035"
11	1	+ .00104"	- .00015"	.0015 "	- .0007"
12	1	+ .0012 "	+ .00064"	- .0033 "	- .0080"

Table XIa  
Core Properties of Manganese-Molybdenum Steel

Steel No.	Treatment No.	Elastic Limit Lbs. Per Sq. In.	Maximum Lbs. Per Sq. In.	% Elong.	% R. A.	Plug	Brinell Surface of 1" Rd.	Izod Ft. Lbs. Brin.
12		57,500	118,500	24.2	65.6	262	255-269	103 248

to decarburization and I took some other samples and reheated them in cyanide and they were still soft. The structure on the outside of these parts was a troostitic structure while the eutectoid zone was martensitic. The samples annealed in an atmosphere of nitrogen showed there was no decarburization.

I would like to say that we are really presenting this paper from a practical standpoint rather than from a scientific standpoint. The steels which we have used have been the ones that are in common use for gears and some of the newer steels that have been frequently mentioned.

Some of this information came rather indirectly to Mr. Strauss, and he may have not received it fully. In regard to the steel hardening in the larger sections in oil we found that the satisfactory hardness was obtained by quenching direct from the carburizing pot into oil on a bar 0.875 inches in di-

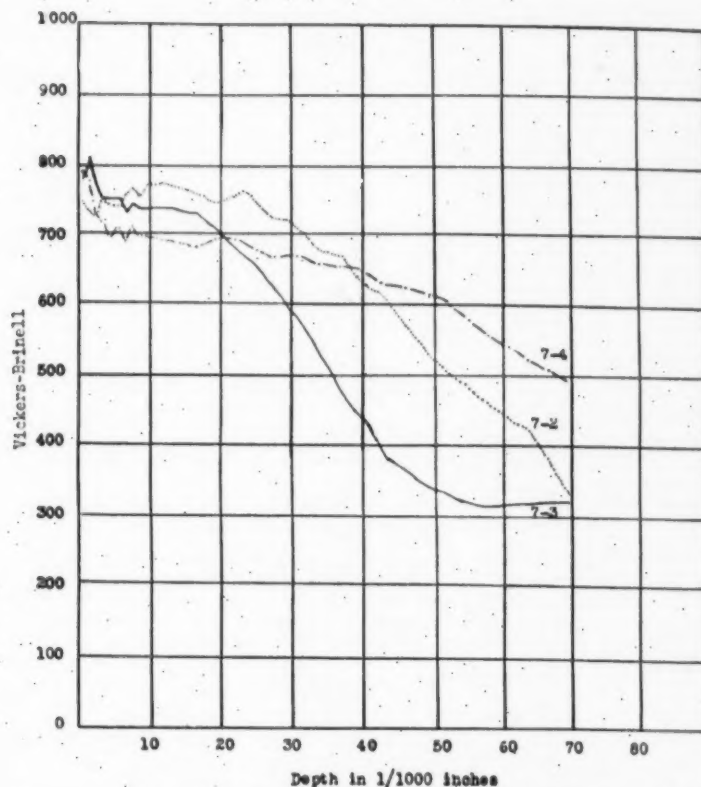


Fig. 9—Depth-hardness Curves of Case Hardened S.A.E. 4615 Steel Carburized 8 Hours (curve 7-3) and 16 Hours (curve 7-4) Cooled in the Carburizing Pot and Reheated to 1525 degrees Fahr. and Oil-quenched. Also S.A.E. 4615 Quenched Direct from the Carburizing Pot (curve 7-2).

ameter whereas these same bars reheated anywhere over a range from 1400 to 1600 degrees Fahr. and quenched in oil did not show a satisfactory hardness. This is contrary to the results we have obtained in other steels. The lower the hardening temperature and still being well above the critical range, of course, the higher the surface hardening obtained. In regard to the finish of the bars which we have tested they were machined and ground, but not polished except from what small amount of polish they had in cleaning off the scale after heat treating which was not sufficient to remove grinding marks. As to the size of the special S.A.E. 6115 steel we used some years ago a great quantity of that steel both in smaller type passenger gears and in much larger truck and bus gears. Of course these gears were quenched with oil under pressure. It is possible that the quenching is more drastic than an agitated quenching by hand in still oil would be. As far as the effect of cyanide on the surface hardness is concerned whether it is a question of decarburization or not, I did a little work on this in examining the pieces at 500 diameters which was the highest we used. On a longer time in cyanide for maybe 0.002 or 0.003 inch below the surface there were a number of little holes like pit holes similar to polishing pits, but these were not observed in the short-time cyanided samples. Just what the cause of that is I do not know.



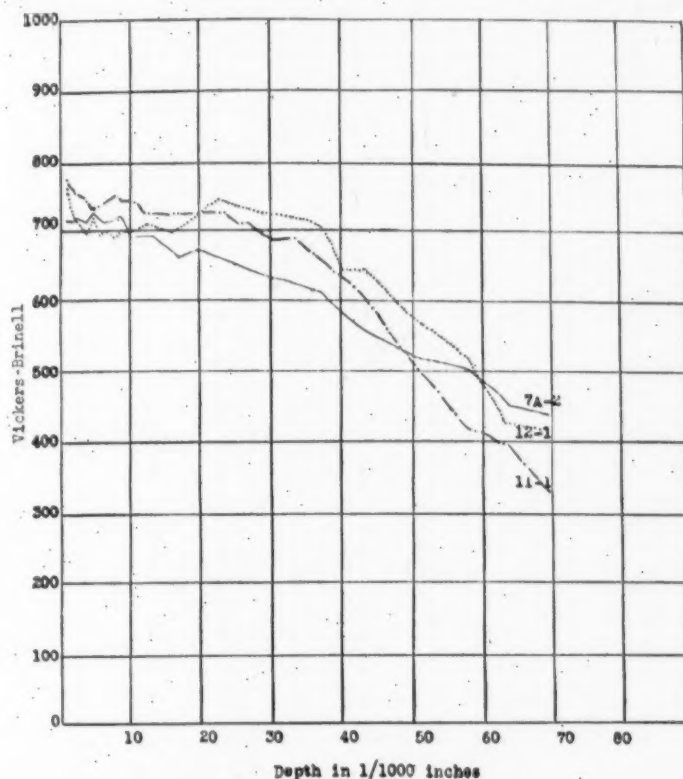


Fig. 10—Depth-hardness Curves of Case Hardened Manganese and Manganese-molybdenum Steels. Also S.A.E. 4615 Steel Oil-quenched from Carburizing Pot.

#### Authors' Written Reply

Some additional work has been done since this paper was originally presented. The work includes tests on shallow and deeply carburized and oil-quenched S.A.E. 4615 steel, and on carburized and oil-quenched manganese and manganese-molybdenum steel; also on fine or coarse-grained single and double water-quenched S.A.E. 1020 steel.

It will be noticed that the deeper case considerably lowers the Izod values for case hardened S.A.E. 4615 but increased the static load by 25 per cent. The manganese and manganese-molybdenum steels show good static load-carrying capacities but rather low impact values in the case hardened condition when compared with some of the other alloy steels. In the uncarburized condition, however, very excellent physical properties were obtained.

The fine-grained S.A.E. 1020 steel shows much better properties than the coarse-grained, both under impact and static loads. Both were improved by double treating, the coarse-grained steel more so than the fine. The impact values indicate that where sharp corners, notches, grooves, etc., are absent fine-grained steel, single treated gives good results, while coarse-grained steel should be double treated to secure better shock resistance. The fine and coarse-grained steels show a difference in character of dimension changes after

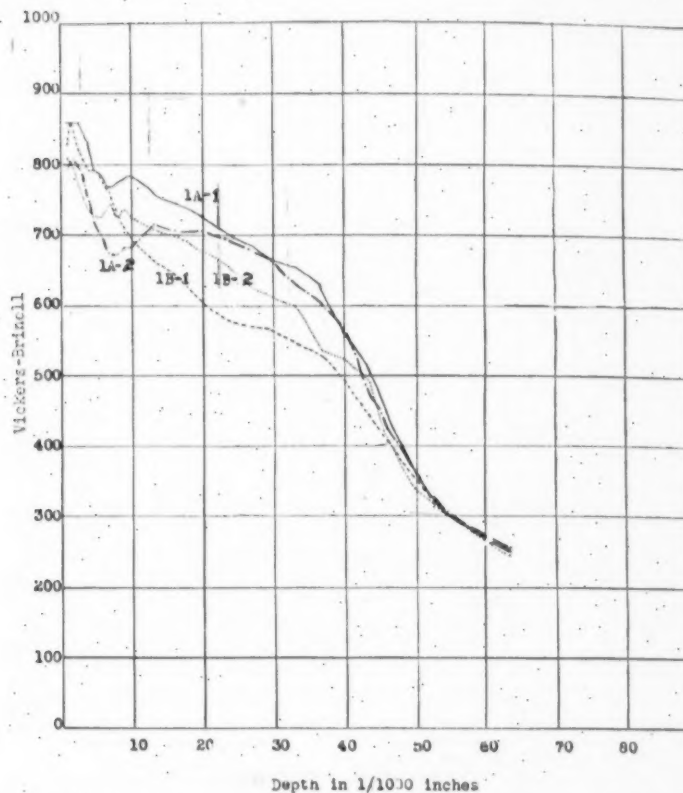


Fig. 11—Depth-hardness Curves of Case Hardened S.A.E. 1020 Steel. All Carburized and Cooled in the Pot. Curve 1A-1 Fine-grained Steel Single-quenched in Water from 1430 degrees Fahr. Curve 1B-1 Coarse-grained Steel Single-quenched in Water from 1430 degrees Fahr. Curve 1A-2 Fine-grained Steel Quenched from 1575 degrees Fahr. into Water, Reheated to 1430 degrees and Water-quenched. Curve 1B-2 same as 1A-2 except Coarse-grained Steel.

carburizing; in general the fine-grained steel showing less change from the original. The fine-grained steel was fine-grained normal and gave hardness values practically equal to those on the coarse-grained steel.

## A STUDY OF BURNING AND OVERHEATING OF STEEL— PART III

BY W. E. JOMINY

### *Abstract*

*This paper shows the effect on the physical properties of steel produced by heating to temperatures approaching but below the burning point of the steel. The study was made in order to determine the maximum temperature to which the steel may be heated in a forging furnace without harming it. The properties of steel heated to these high temperatures and subsequently annealed or heat treated are compared with the corresponding properties of the same steel not heated previous to annealing or heat treating. The steels studied include S.A.E. Nos. 1015, 1030, 1050, 1085, and 3145. A table is given showing the maximum temperature at which these steels may be heated if overheating and burning are to be avoided.*

**L**IMITING temperature for some of the common forging steels below which forging can be accomplished without danger of burning are recorded in Parts I and II of this discussion which deal with the general subject of the burning of steel.<sup>1</sup> There has been some question, however, whether these steels, although not actually burned, will in any way be harmed by being heated to these comparatively high temperatures just under the burning point. Some metallurgists have claimed that forging may be satisfactorily done at these high temperatures if the forging is continued until the metal has become relatively cool; but, if, the forging is discontinued while the metal is still relatively hot, they claim that the properties of the steel will be impaired and cannot be restored by ordinary heat treatment. Steel in this condition is sometimes spoken of as being overheated. The term "overheated" is used to describe so many phenomena that it may prove misleading unless used in connection with

<sup>1</sup>W. E. Jominy, "A Study of Burning and Overheating of Steel" Parts I and II, TRANSACTIONS, American Society for Steel Treating, Vol. 16, August and September 1929, pp. 298 and 372.

This paper constitutes part of an investigation carried out for the American Gas Association on the subject of forging, at the University of Michigan.

A paper presented before the Eleventh Annual Convention of the Society, Cleveland, September 9 to 13, 1929. The author, W. E. Jominy, member of the society, is associated with the Engineering Research department of the University of Michigan, Ann Arbor, Michigan. Manuscript received August 22, 1929.

the related term "burned." An overheated steel is differentiated from a burned steel in that the latter contains intercrystalline voids, whereas, steel which is merely overheated (or perhaps a less confusing term would be "excessively heated") does not contain these.

It is to be expected that if steel is heated to very high temperatures and not subsequently annealed or heat treated, that its physical properties will be impaired, and, therefore, it is generally accepted that, if steel is not subsequently heat treated, its physical properties will depend in part on the temperature at which the forging was stopped.

In this paper we are concerned only with steel which is subsequently heat treated after forging and whose physical properties will then depend in part on the heat treatment which follows the forging. The question arises, if the forging of steel is discontinued at a very high temperature and the steel is then properly normalized, will its physical properties differ from those of another sample of the same steel which has been forged at a lower temperature and similarly normalized? L. S. Cope<sup>2</sup> has reported that heating to a high temperature without sufficient subsequent forging will result in large grain size which even freak methods of heat treatment will only partially refine, but which proper forging will entirely eliminate. This same phenomenon is also reported by O. A. Knight.<sup>3</sup> D. K. Bullens<sup>4</sup> has presented data showing the poorer physical properties, fractures, and microstructures of steels that have been heated too hot in the forging operations after they have all been given the same subsequent heat treatment to refine the steel. Although these metallurgists and others report that steel is harmed by this manner of heating, they fail to give the temperatures at which the phenomena occurs. It is thus impossible to learn from their work at what temperature the forging should be done in order to eliminate any possibility of harming the steel.

The object of the work covered by this paper has been to determine whether heating steel to temperatures approaching, but below the burning point, will affect its physical properties so that these effects will not be removed by subsequent heat treatment.

<sup>2</sup>L. S. Cope, "Coarse-Grained Drop Forgings—Their Detection and Correction," *TRANSACTIONS, American Society for Steel Treating*, Vol. 3, 1923, p. 808.

<sup>3</sup>O. A. Knight, "Defective Forgings," *Iron Age*, Vol. 103, 1919, p. 1696.

<sup>4</sup>D. K. Bullens, "Forging Versus Heat Treatment of Steel," *Iron Age*, Vol. 99, 1917, p. 1243.



Further, if it were found that the physical properties of steel were impaired by this high temperature exposure, the investigation aimed to determine at what temperature the steel is injured in order to find a maximum temperature at which steel may be forged without danger of its being harmed.

Since it is accepted that forging until the temperature is sufficiently lowered will eliminate any harmful effects of high temperature heating, it is the temperature at which forging is discontinued that determines whether or not the steel will be harmed by being overheated. This condition differs from that of burning since the significant temperature in the latter case is that of the steel previous to and at the time of starting the forging operation. It is difficult to accurately measure the temperature of a steel at the time forging is discontinued. But, if a steel is reheated to a given temperature higher than that at which the forging was discontinued, there should be the same effect as would result from discontinuing the forging at that temperature.

#### PROCEDURE AND RESULTS

In accordance with the view just expressed, the general procedure followed in this study was to heat the steel previously rolled to  $\frac{3}{4}$  inches round to temperatures approaching the burning point, and then, after suitable annealing or heat treating, to test it. The results so obtained were compared with the results obtained from tests made on bars of the same steel which were not heated to these high temperatures but which were given the same anneal or heat treatment. In studying any one steel, the test bars were made, insofar as possible, by cutting them all off the same long bar of steel. When more test bars were needed than could be made from one of these bars of steel a second bar was used from what was believed to be the same heat of steel and, consequently, whose chemical composition was nearly the same. In some cases a third and fourth such long bar were cut up for test specimens. In all such cases specimens from an unheated test bar were used for comparison with the test specimens cut from each long bar of steel. All bars were heated in the gas-fired muffle furnace described in Part I of this discussion.

Tests were made by heating bars 6 inches long and  $\frac{3}{4}$  inches round in lots of three to various temperatures. The highest of these was just below the burning point of the particular steel, and the others were successively lower by intervals of from 25 to 100 degrees

Table I  
Chemical Composition of Steels Tested

Steel S. A. E. No.	Per Cent Carbon	Per Cent Manganese	Per Cent Sulphur	Per Cent Phosphorus	Per Cent Nickel	Per Cent Chromium
1015	0.17	0.58	0.030	0.017	.....	.....
1030	0.30	0.65	0.033	0.012	None	Trace
1050	0.52	0.64	0.034	0.025	None	None
1085	0.86	0.26	0.018	0.029	None	None
1085	0.88	0.34	0.018	0.029	None	None
1095	0.99	0.31	0.035	0.020	None	None
3145	0.45	0.61	0.019	0.018	1.27	0.53
3145	0.44	0.64	0.018	0.011	1.20	0.74
3145	0.38	0.53	0.021	0.028	1.27	0.61

Fahr., depending on individual conditions, until temperatures were reached at which there appeared little danger of harming the steel. In one series of tests all the bars thus heated, and also a few not heated at all, were annealed at the same time in the same furnace. In another series of tests a heat treatment consisting of annealing, hardening, and tempering was used after the high temperature treatment in an effort to determine to what extent this treatment might affect steels that would subsequently be hardened. After being given these treatments, the bars were subjected to hardness and tensile tests. It was thought that in some cases the steel might not show any difference in tensile properties due to high temperature exposure but might have other properties affected. Consequently, impact tests, McQuaid-Ehn tests, electrical conductivity tests, microscopic examinations, and also macroscopic examinations were made on some of the test bars.

Five steels were used in this investigation. These included four plain carbon steels and one nickel-chromium steel. The four plain carbon steels consisted of an S.A.E. 1015 steel, an S.A.E. 1030 steel, an S.A.E. 1050 steel, and an S.A.E. 1085 steel. The alloy steel was an S.A.E. 3145 steel. Table I shows the chemical composition of the steels investigated.

*Tests on S.A.E. 1015 Steel.* A summary of the results of the tests made on the S.A.E. 1015 steel are reported in Table II.

In making these tests the bars were held for ten minutes at the temperature given in the table, the total time in the furnace varying from 20 to 30 minutes. After being held for ten minutes at heat, the furnace was turned off, the door opened, and the piece allowed to cool 200 degrees Fahr. The bars were then removed and cooled in sand. A comparison of the data derived from specimens 315-A, B, and C with that from specimens 319-A, B, and C, will indicate

Table II  
Summary of Tensile, Hardness and McQuaid-Ehn Tests on S.A.E. 1015 Steel

Specimen	Heat Treatment Deg. F.	Yield Point #/in. <sup>2</sup>	Tensile Strength #/in. <sup>2</sup>	Breaking Load #/in. <sup>2</sup>	% Elongation	% Reduction of Area	Sclero-scope Hardness	B Scale Rockwell Hardness	Brinell Hardness	Ehn Grain Size
315 A	2675	30,500	55,500	37,250	41.25	67.9	12	56	107	7
B	2675	30,250	63,500	47,700	41.0	69.0	13	57	116	7
C	2675	30,000	65,300	50,500	41.5	68.8	..	..	..	..
Av.	2675	30,250	61,430	45,150	41.25	68.6	13	57	112	..
316 A	2375	29,000	54,500	36,000	43.0	69.3	14	56	112	6
B	2375	29,000	54,900	35,550	41.5	70.4	12	57	107	6
C	2375	28,750	54,750	36,475	42.5	68.6	..	56	..	..
Av.	2375	28,916	54,717	36,008	42.3	69.4	13	56.5	109	..
317 A	2475	28,500	54,500	36,500	41.0	68.6	14	57	114	6-7
B	2475	28,000	54,500	36,000	41.5	68.6	13	55	112	6
C	2475	29,000	54,750	36,700	42.0	68.6	..	..	..	..
Av.	2475	28,500	54,583	36,400	41.5	68.6	13.5	56	113	..
318 A	2375	27,500	54,800	37,000	41.75	67.9	14	61	107	6
B	2375	27,500	54,800	36,500	41.5	68.6	13	55	114	6
C	2375	28,250	54,400	36,500	42.0	67.7	..	..	..	..
Av.	2375	27,750	54,667	36,667	41.75	68.1	13.5	58	110.5	..
319 A	None	28,500	55,400	37,750	40.0	66.3	13	57	109	6
B	None	28,000	55,000	37,800	40.0	66.5	13	56	109	6
C	None	28,000	55,050	37,250	40.0	66.5	..	..	..	..
Av.	None	28,167	55,150	37,600	40.0	66.4	13	56.5	109	..

## Chemical Analysis of the Steel:

Carbon	0.17 %
Manganese	0.58 %
Sulphur	0.030 %
Phosphorus	0.007 %

All specimens were annealed for 40 minutes at 1625 degrees Fahr.

that no appreciable difference in properties exist between the bars which were heated to the very high temperature of 2675 degrees Fahr. and subsequently annealed, and those that were not heated at all previous to the annealing, except that for some inexplicable reason the tensile strength and the breaking load of the bars heated to 2675 degrees Fahr. averaged somewhat higher than for the bars not heated. No undesirable effect from overheating is shown for this steel, provided of course it is not burned.

A section of specimen 315A, which had been heated to 2675 degrees Fahr. and annealed, was polished, etched with four per cent nitric acid in alcohol, and examined under the microscope. There was no evidence of burning or persistent gamma-iron grain boundaries or other abnormal condition. Photomicrograph Fig. 1 shows the structure of this specimen.

Photomicrograph of Fig. 2 was made of specimen 319A, which is of the same steel as 315A, and was annealed in exactly the same manner, but was not heated previous to the annealing. As will be observed in comparing these photomicrographs, there are no observable differences between them by which the sample heated to 2675 degrees Fahr. may be distinguished from the sample not heated.

*Tests on S.A.E. 1030 Steel.* The results obtained on the S.A.E. 1030 steel as shown in Table III indicate that, after this steel had been properly annealed, no undesirable effects of the previous high temperature treatment were evident.

A section of specimen 342B, which was heated to 2625 degrees Fahr. and annealed at 1600 degrees Fahr., was polished, etched, and examined under the microscope. There was nothing abnormal in the microstructure of this steel, the structure being quite similar to that of the specimens which were not heated previous to annealing. Fig. 3 shows the microstructure of specimen 342B, and Fig. 4 shows the same for specimen 347A. The latter was not heated previous to being annealed at 1600 degrees Fahr.

*Tests on S.A.E. 1050 Steel.* A series of tests was made with an S.A.E. 1050 steel to investigate the effects of heating within the temperature range from 2500 to 2600 degrees Fahr. In this series of tests, bars in sets of three were heated to 2600, 2575, 2550, 2525 and 2500 degrees Fahr., respectively. These were all maintained for the standard length of time of ten minutes at heat, and were in the furnace a total time of from 20 to 25 minutes. Another set of three test bars was maintained at 2500 degrees Fahr. for two hours, and



Table III  
Summary of Tensile and Hardness Tests on S.A.E. 1030 Steel

Specimen	Heat Treatment Deg. F.	Annealed 40 Min. Deg. F.	Yield Point #/in. <sup>2</sup>	Tensile Strength #/in. <sup>2</sup>	Breaking Load #/in. <sup>2</sup>	% Elongation	% Reduction of Area	Sclero- scope Hardness	B Scale Rockwell Hardness	Brinell Hardness
342 A	2625	1600	39,910	66,145	52,300	36.5	52.0	16	66	126
B	2625	1600	43,700	66,100	51,900	35.5	56.8	17	67	127
C	2625	1600	39,300	66,100	52,200	36.0	56.5			
Av.	2625	1600	40,705	66,115	52,130	36.0	55.1	16.5	66.5	126
343 A	2600	1600	39,250	65,600	50,550	38.0	58.3	16	67	119
B	2600	1600	40,950	65,600	51,800	36.0	57.0	16	66	120
C	2600	1600	36,750	65,500	50,600	37.0	58.1			
Av.	2600	1600	38,985	65,570	50,985	37.0	57.8	16	66.5	119
344 A	2575	1600	38,500	65,710	51,335	37.0	57.1	16	66	109
B	2575	1600	42,345	65,575	50,890	36.5	58.0	16	65	112
C	2575	1600	38,600	65,950	50,850	38.5	58.1			
Av.	2575	1600	39,815	65,745	51,030	37.3	57.7	16	65.5	110
345 A	2550	1600	43,000	65,700	51,700	37.5	56.8	16	65	118
B	2550	1600	36,750	65,450	51,600	38.0	56.5	16	66	121
C	2550	1600	39,600	65,500	52,150	38.0	56.5			
Av.	2550	1600	39,785	65,550	51,820	37.8	56.6	16	65	119
346 A	2525	1600	40,750	64,910	50,890	38.5	57.7	16	65	114
B	2525	1600	39,810	65,600	51,600	35.5	56.9	16	65	114
C	2525	1600	38,500	65,400	50,850	37.0	57.3			
Av.	2525	1600	39,690	65,305	51,115	37.0	57.3	16	65	114
347 A	None	1600	41,650	64,160	50,600	37.0	57.7	17	63	112
B	None	1600	42,200	64,550	51,200	37.0	56.2	17	63	112
C	None	1600	39,000	64,700	51,000	38.0	56.8			
Av.	None	1600	40,950	64,470	50,935	37.3	56.9	17	63	112

## Chemical Analysis of the Steel:

Carbon	0.30 %
Manganese	0.65 %
Chromium	Trace
Nickel	None
Sulphur	0.033%
Phosphorus	0.012%

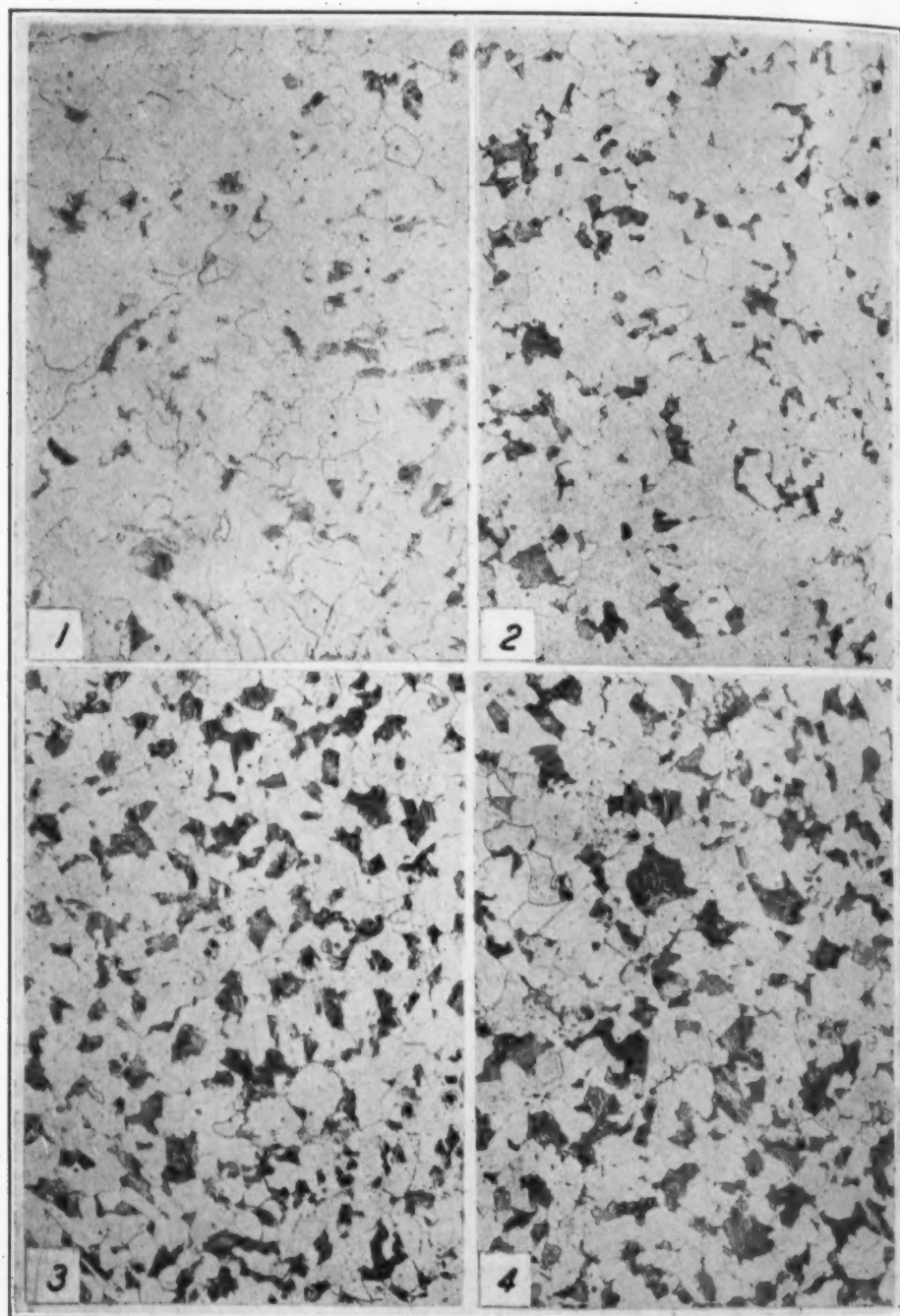


Fig. 1—Specimen 315A, a Sample of S. A. E. 1015 Steel Heated to 2675 Degrees Fahr., and Annealed at 1625 Degrees Fahr.

Fig. 2—Specimen 319A, a Sample of S. A. E. 1015 Steel not Heated Previous to Annealing at 1625 Degrees Fahr.

Fig. 3—Specimen 342B, a Sample of S. A. E. 1030 Steel Heated to 2625 Degrees Fahr. and Annealed at 1600 Degrees Fahr.

Fig. 4—Specimen 347A, a Sample of S. A. E. 1030 Steel not Heated Before Annealing at 1600 Degrees Fahr.

All Specimens Etched with 4 Per Cent Nitric Acid in Alcohol. All Magnifications  $\times 150$ .

a third set was not heated at all. These bars were then all annealed at 1575 degrees Fahr. in the same furnace at the same time and machined into tensile bars and tested. Table IV gives a summary of the results obtained in this series of tests. It will be noted in examining this table that there is a sharp break in the physical properties of the test bars between the 2550 and 2575 degrees Fahr. heats. The bars heated to 2575 degrees Fahr. and higher and subsequently annealed are more brittle than are the bars heated to 2550 degrees Fahr. or below. This is shown by the lower per cent elongation and reduction in area and by the higher breaking load of the bars heated to 2575 degrees Fahr. and above.

Specimen 355A, which was heated to 2575 degrees Fahr., and then subjected to a metallographic examination, and was found to show no evidence of being burned. Nothing could be found in the microstructure of the steel to account for its relatively poor ductility as shown in Table IV. However, it is possible that the overheated or burned areas were segregated and that the sections polished did not happen to contain one of these areas. Fig. 5 shows the structure of this steel.

Specimen 356B, which was heated to 2550 degrees Fahr. and annealed, appeared to have a structure quite similar to that of specimen 355A, although, as may be observed from Table IV, specimen 356B, had better ductility. The microstructure of this specimen is shown in Fig. 6.

After these tests had been made, it was felt that perhaps if the steel were in the hardened and tempered condition rather than in the annealed condition, defects or changes due to overheating might be more evident. Consequently, test bars of the S.A.E. 1050 steel were heated to 2550, 2525, 2500, and 2450 degrees Fahr., three test bars being used for each heat. The manner of heating and the time at heat were the same as in the previous heats already described. These bars together with three bars which had not been heated were placed in a furnace and annealed by heating to 1575 degrees Fahr. for a total time at heat of 40 minutes, and were then cooled in air. They were then reheated to 1500 degrees Fahr., held at heat for 40 minutes, and quenched in water. Following this treatment they were heated to 1020 degrees Fahr. for one hour. The bars were machined into tensile specimens, and tested for tensile properties and hardness. Table V gives a summary of the results of these tests. It will be observed that the bars heated to 2550 degrees Fahr. are more brittle

Table IV  
Summary of Tensile, Hardness and McQuaid-Ehn Tests on S.A.E. 1050 Steel

Specimen	Heat Treatment Deg. F.	Annealed 40 Min. Deg. F.	Yield Point #/in. <sup>2</sup>	Tensile Strength #/in. <sup>2</sup>	Breaking Load #/in. <sup>2</sup>	% Elongation	% Reduction of Area	Sclero- scope Hardness	B Scale Rockwell Hardness	Brinell Hardness	Ehn Grain Size
336 A	2600	1575	48,750	84,350	84,350	12.0	14.1	20	82	156	6-7
B	2600	1575	49,000	84,200	83,250	15.5	15.6	20	82	149	..
C	2600	1575	48,750	83,450	83,450	12.0	14.5	..	..	..	..
Av.	2600	1575	48,830	84,000	83,650	13.1	14.7	20	82	152	..
334 A	2600	1575	49,050	85,000	85,000	13.5	10.7	22	80	155	..
B	2600	1575	52,300	85,000	83,700	13.5	11.1	22	81	153	..
C	2600	1575	52,275	84,703	83,950	13.5	11.0	..	..	..	..
Av.	2600	1575	51,210	84,900	84,215	13.5	10.9	22	81	154	..
335 A	2575	1575	46,850	84,950	84,950	16.0	16.6	22	83	159	..
B	2575	1575	48,507	84,640	83,745	16.0	17.0	22	82	155	..
C	2575	1575	48,200	84,700	83,500	16.0	18.1	..	..	..	..
Av.	2575	1575	47,850	84,765	84,065	16.0	17.3	22	82.5	157	..
336 A	2550	1575	49,750	85,300	76,500	26.5	40.1	23	82	156	..
B	2550	1575	47,800	85,050	77,050	26.5	40.4	22	82	155	..
C	2550	1575	50,050	85,250	75,750	27.5	40.7	..	..	..	..
Av.	2550	1575	49,200	85,200	76,435	26.8	40.4	22.5	82	156	..
335 A	2525	1575	45,325	85,650	75,595	26.0	40.3	22	82	146	6-7
B	2525	1575	47,700	85,800	75,700	26.5	41.6	22	82	163	..
C	2525	1575	47,820	85,515	76,130	27.0	42.5	..	..	..	..
Av.	2525	1575	46,950	85,655	75,810	26.5	41.5	22	82	154	..
334 A	2500	1575	43,850	85,250	76,900	26.5	38.2	22	82	143	5
B	2500	1575	44,800	85,450	76,350	27.0	40.1	22	82	149	..
C	2500	1575	47,500	85,750	77,850	26.0	39.4	..	..	..	..
Av.	2500	1575	45,180	85,470	77,030	26.5	39.2	22	82	146	..
338 A	2500	1575	48,500	80,350	71,850	27.5	43.2	19	80	151	5
B	2 hrs.	1575	48,000	81,500	72,850	28.5	43.4	20	79	149	..
C	2 hrs.	1575	47,350	80,700	71,700	26.5	43.4	..	..	..	..
Av.	2 hrs.	1575	47,220	80,850	72,130	27.5	43.3	19.5	80	150	..
337 A	None	1575	49,250	87,100	77,750	26.0	37.0	20	81	146	6
B	None	1575	45,350	87,200	77,750	26.0	36.5	22	82	153	..
C	None	1575	47,000	87,300	79,200	26.5	37.2	..	..	..	..
Av.	None	1575	47,200	87,200	78,230	26.2	36.9	21	81	150	..

Chemical Analysis of the Steel:

Carbon 0.52%, Manganese 0.64%, Sulphur 0.034%, Phosphorus 0.025%



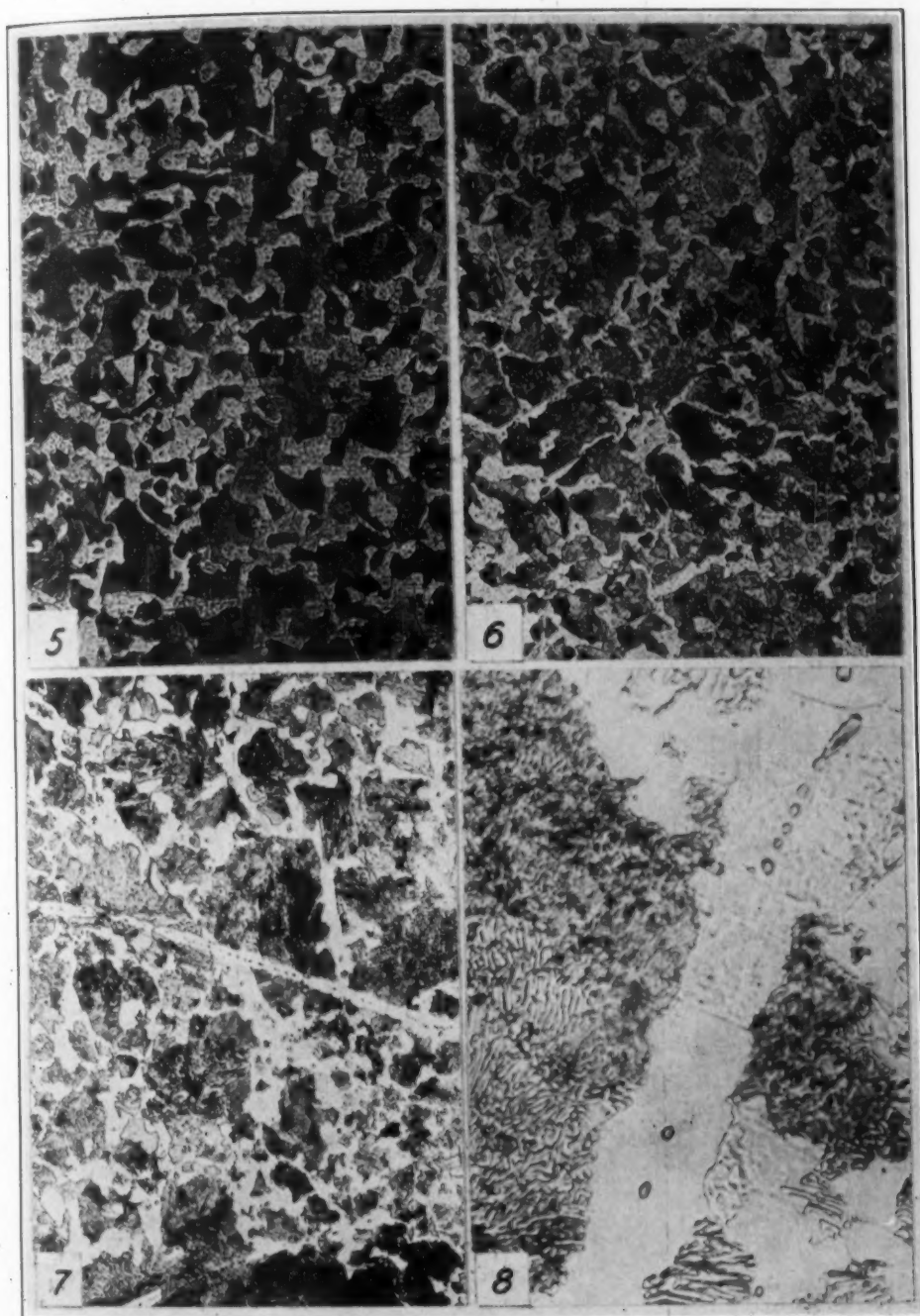


Fig. 5—Specimen 355A, a sample of S. A. E. 1050 Steel, Heated to 2575 Degrees Fahr., and Annealed at 1575 Degrees Fahr. Magnification  $\times 150$ .

Fig. 6—Specimen 356B, a Sample of S. A. E. 1050 Steel Heated to 2550 Degrees Fahr., and Annealed at 1575 Degrees Fahr. Magnification  $\times 150$ .

Fig. 7—Specimen 376A, a Sample of S. A. E. 1050 Steel, Heated to 2550 Degrees Fahr., Annealed at 1575 Degrees Fahr., Quenched into Water from 1500 Degrees Fahr., Tempered at 1020 Degrees Fahr., and Annealed Again at 1575 Degrees Fahr. Magnification  $\times 150$ .

Fig. 8—Same Specimen as Fig. 7. Note the Nonmetallic Inclusions at the Grain Boundaries. Magnification  $\times 1250$ .

All Specimens Etched with 4 Per Cent Nitric Acid in Alcohol.

Table V  
Summary of Tensile and Hardness Tests on S.A.E. 1050 Steel—Heated, Quenched and Tempered

Specimen	Heat Treatment	Anneal	Quench (water) Deg. F.	Temper (1 hr.) Deg. F.	Yield Point Lbs./in. <sup>2</sup>	Tensile Strength Lbs./in. <sup>2</sup>	Breaking Load Lbs./in. <sup>2</sup>	Per Cent Elongation in 2 in. of Area	Per Cent Reduction of Area	Brinell	Hardness Rockwell C	Sclerometer
376	A	1575°F.	1500	1020	89,250	124,400	121,350	9.0	10.4	223	C24	30
	H	1575°F.	1500	1020	89,150	123,350	123,350	6.0	10.0	228	C24	34
	C	1575°F.	1500	1020	90,000	124,150	123,350	9.0	12.6	225	C24	32
	Av.	1575°F.	1500	1020	89,465	123,965	122,685	8.0	11.0	225	C24	32
389	A	1575°F.	1500	1020	86,650	121,550	86,950	21.2	55.7	241	C19	30
	B	1575°F.	1500	1020	86,750	119,750	85,550	20.5	55.7	239	C20	30
	C	1575°F.	1500	1020	88,000	121,200	86,500	21.3	55.4	240	C19.5	30
	Av.	1575°F.	1500	1020	87,135	120,835	86,335	21.0	55.6	240	C19.5	30
382	A	1575°F.	1500	1020	87,850	123,440	88,650	19.0	54.9	217	C24	33
	B	1575°F.	1500	1020	88,750	121,500	87,750	20.0	56.2	228	C24	34
	C	1575°F.	1500	1020	86,850	122,400	91,550	19.5	54.1	223	C24	33
	Av.	1575°F.	1500	1020	87,820	122,450	89,320	19.5	55.1	223	C24	33
383	A	1575°F.	1500	1020	87,900	126,850	91,300	17.0	54.9	252	C27	31
	B	1575°F.	1500	1020	93,600	124,850	92,400	17.0	53.6	252	C21	31
	C	1575°F.	1500	1020	92,900	124,650	88,800	19.0	55.2	252	C24	31
	Av.	1575°F.	1500	1020	91,470	125,450	90,830	17.7	54.6	252	C24	31
377	A	1575°F.	1500	1020	93,710	127,250	96,410	18.0	50.7	230	C24	32
	B	1575°F.	1500	1020	92,750	127,900	97,750	18.5	50.0	226	C25	30
	C	1575°F.	1500	1020	95,500	131,250	102,350	19.0	46.9	228	C24	31
	Av.	1575°F.	1500	1020	93,987	128,800	98,835	18.5	49.2	228	C24	31

## Chemical Analysis of the Steel:

Carbon	0.52 %
Manganese	0.64 %
Sulphur	0.034 %
Phosphorus	0.025 %

than are the bars which were heated to 2525 degrees Fahr. or temperatures below. This is indicated by the decided difference in per cent elongation and reduction of area of specimens 376A, B, and C as compared with these qualities of specimens 389A, B, and C.

An examination was made of the microstructure of some of these samples, and it was found that they appeared to have similar structures. The structure in the hardened condition, however, was so fine-grained that it was decided to anneal the samples and to re-examine them. They were, therefore, annealed at 1575 degrees Fahr. in a partial vacuum for 40 minutes. The samples were again polished, etched, and examined under the microscope. This time the sample heated to 2550 degrees Fahr. showed large persistent grain boundaries. At these grain boundaries were grains of ferrite that did not appear to migrate or absorb carbon during the heat treatments following the high temperature heating. At the center of these ferrite grains there were nonmetallic inclusions. These are shown in Fig. 7. Fig. 8 shows this structure at 1250 diameters. Specimens heated to 2525 degrees Fahr. and below did not show this type of structure and appeared to be quite normal.

The structure displayed in Figs. 7 and 8 has been observed before in samples heated to the burning point. Such heating appears to cause segregation and precipitation of nonmetallic inclusions in the grain boundaries. Probably due to the relatively high percentage of these inclusions in the ferrite the latter does not readily dissolve carbon or behave as normal ferrite. Consequently, further heat treatment does not break up this structure and the planes of weakness remain. This appears to be the first stage of burning.

A series of impact tests were made in order to study further the behavior of this steel when heated to high temperatures. Three test bars of this steel labeled A, B, and C, were accordingly heated to 2575 and 2550 degrees Fahr. for a total time in the furnace of about 30 minutes. Three additional bars were heated to 2500 degrees Fahr. for a total time of two hours and 18 minutes in the furnace, and three bars were not heated. One bar each of this series, the bars labeled A, were heated to 1575 degrees Fahr. in the same furnace at the same time, and cooled in the furnace. The bars were then each cut into three impact test bars and tested in the Izod impact testing machine. The summary of results of this test are given in Table VI. Examination of the table shows that there is little difference in the impact resistance of these bars.

**Table VI**  
**Summary of Izod Impact Values for S.A.E. 1050 Steel**

Specimen	Temperature to which steel was heated Deg. F.	Time in Furnace (minutes)	Time at Heat (minutes)	Subsequent Heat Treatment	Impact Values (Ft. Lbs.)
368 A1	2575	30	10	Annealed at 1575°F.	7.5
A2	2575	30	10		6.0
A3	2575	30	10		7.5
Av.	2575	30	10		7.0
369 A1	2550	31½	8½	1575°F.	6.0
A2	2550	31½	8½		6.0
A3	2550	31½	8½		7.5
Av.	2550	31½	8½		6.5
370 A1	2500	2 hrs.-18 min.	2 hrs.	1575°F.	6.5
A2	2500	2 hrs.-18 min.	2 hrs.		6.5
A3	2500	2 hrs.-18 min.	2 hrs.		10.0
Av.	2500	2 hrs.-18 min.	2 hrs.		7.7
407 A1	Not heated	.....	.....	1575°F.	7.5
A2	Not heated	.....	.....		11.0
A3	Not heated	.....	.....		5.5
Av.	Not heated	.....	.....		8.0

Chemical Analysis of the Steel:

Carbon	0.53 %
Manganese	0.64 %
Sulphur	0.034%
Phosphorus	0.025%

One each of the bars in the series described above the bars labeled B, were treated by being annealed at 1575 degrees Fahr., then quenched in oil from 1700 degrees Fahr. and finally reannealed at 1575 degrees Fahr. In these treatments all the bars were heated in the same furnace at the same time. The results of these tests are shown in Table VII. Here again no differences in resistance to impact attributable to overheating could be found.

The bars labeled C in this series were quenched from the overheating temperature directly into oil and were all subsequently annealed at 1575 degrees Fahr. This of course is not a commercial treatment but it may happen to be given to an occasional forging. The results of these tests, which are given in Table VIII, show slightly lower impact values for bars heated to the higher temperatures.

The impact tests on this steel seems to indicate that the steel was little affected by the high temperature treatments; in fact, only when the steel was quenched from the temperature of overheating was any effect apparent. In that case the bars heated to the highest temperatures had the least resistance to impact after annealing.

It was felt that perhaps electrical conductivity tests might show some difference between excessively heated steel and steel not exces-



Table VII  
Summary of Izod Impact Values for S.A.E. 1050 Steel

Specimen	Temperature to which steel was heated Deg. F.	Time in Furnace (minutes)	Time at Heat (minutes)	Subsequent Heat Treatment	Impact Values (Ft. Lbs.)
368 B1	2575	30	10	All specimens annealed at 1575°F., oil quenched from 1700°F. and again annealed at 1575°F.	6.5
B2	2575	30	10		5.0
B3	2575	30	10		7.5
Av.	2575	30	10		6.3
369 B1	2550	31½	8½		6.5
B2	2550	31½	8½		6.5
B3	2550	31½	8½		6.5
Av.	2550	31½	8½		6.5
370 B1	2500	2 hrs.-18 min.	2 hrs.		8.0
B2	2500	2 hrs.-18 min.	2 hrs.		6.5
B3	2500	2 hrs.-18 min.	2 hrs.		8.5
Av.	2500	2 hrs.-18 min.	2 hrs.		7.7
408 B1	Not heated	.....	.....		6.0
B2	Not heated	.....	.....		4.5
B3	Not heated	.....	.....		6.0
Av.	Not heated	.....	.....		5.5

Chemical Analysis of the Steel:

Carbon	0.53 %
Manganese	0.64 %
Sulphur	0.034 %
Phosphorus	0.025 %

sively heated. Accordingly, three test specimens of the S.A.E. 1050 steel were cut from the same long bar and one of the specimens was heated to 2620 degrees Fahr., one to 2575 degrees Fahr., and one was not heated at all. They were all annealed at 1575 degrees Fahr., in the same furnace at the same time. The electrical conductivity of the bars was measured by means of a double Kelvin bridge and a galvanometer. Table IX shows the results obtained in these tests. It will be observed that there is little change in the resistivity of this steel even after severe overheating, provided of course that the bars are subsequently annealed.

*Tests on S.A.E. 1085 Steel.* The original tests on the S.A.E. 1085 steel were in two series. In one series the steel bars were heated to 2550, 2450, 2300 and 2200 degrees Fahr. and annealed and their tensile properties compared with those of bars which were not heated except in the annealing. In the other series of tests the steel bars were heated to 2400, 2385, 2350, 2320, 2290 and 1900 degrees Fahr. and annealed and a comparison made of their tensile properties. The results of the first series of tests are shown in Table X and those of the second series in Table XI. A study of the results given in these tables shows that there are only small changes in the properties of the test bars which were heated within the range from 2200 to 2400 de-

**Table VIII**  
**Summary of Izod Impact Values for S.A.E. 1050 Steel**

Specimen	Temperature to which steel was heated Deg. F.	Time in Furnace (minutes)	Time at Heat (minutes)	Subsequent Heat Treatment	Impact Values (Ft. Lbs.)
368 C1	2575	30	10	All specimens quenched in oil from the temperature heated, and an-	6.0
C2	2575	30	10		6.5
C3	2575	30	10		7.0
Av.	2575	30	10		6.5
369 C1	2550	31½	8½	annealed at 1575° F.	8.5
C2	2550	31½	8½		7.0
C3	2550	31½	8½		10.0
Av.	2550	31½	8½		8.5
370 C1	2500	2 hrs.-18 min.	2 hrs.		10.0
C2	2500	2 hrs.-18 min.	2 hrs.		10.0
C3	2500	2 hrs.-18 min.	2 hrs.		12.5
Av.	2500	2 hrs.-18 min.	2 hrs.		10.8
Chemical Analysis of the Steel:					
Carbon .....				0.53 %	
Manganese .....				0.64 %	
Sulphur .....				0.034%	
Phosphorus .....				0.025%	

degrees Fahr. after annealing. Above 2400 degrees Fahr. the bars become more brittle. The only bars heated below 2200 degrees Fahr. are those heated to 1900 degrees Fahr. and, of course, those not heated at all. All of these latter bars seem to have increased ductility as is indicated by their higher per cent elongation and reduction of area.

It was felt that more tests should be made at temperatures of 2200 degrees Fahr. and below. Accordingly, a series of tests was made in which bars of this steel were heated to 2300, 2200, 2100, 2000, 1895 and 1725 degrees Fahr. and then all annealed at 1500 degrees Fahr. Tensile tests were then made on these bars the results of which are presented in Table XII. In making this test a bar of steel having a somewhat higher carbon content, i. e., 0.99 per cent carbon, was used. The results shown in this table, therefore, are not entirely comparable with those in the previous tables. There is in this case, however, no trend toward greater per cent elongation and greater reduction of area as the initial heating temperature is lowered from 2200 to 1900 degrees Fahr. There does, however, seem to be an increase in these properties when the temperature of heating was lowered to 1725 degrees Fahr. It was suspected that at these relatively lower temperatures the granular or spheroidal cementite present in the original steel might not be completely broken up. Microscopic examination of some of the sections showed this to be the

case. It is, therefore, easily possible that some variation in ductility was due to the presence at the point of fracture of varying proportions of spheroidal and lamellar cementite.

Two additional series of tests were made to show the effect of the presence of the spheroidal cementite as compared with that of the lamellar cementite. In the first series three bars were heated to 2100 degrees Fahr., three to 2000 degrees Fahr., and three to 1700—1750 degrees Fahr. Then these bars were all annealed in such a manner as to completely break up the spheroidal cementite. Following this, tensile tests were made on the bars. A second series of tests were made in which three bars each were heated to the same

Table IX  
Resistivity of S.A.E. 1050 Steel

Specimen	Temperature to which steel was heated °F.	Time in Furnace	Time at Temperature	Annealing Temperature °F.	Electrical resistance ohms	Resistivity microm per square centimeter per centimeter
331	2620	23	9	1575	0.00016056	15.99
332	2575	27	9	1575	0.00015626	15.71
N. H.	Not heated	..	..	1575	0.00015890	15.93

temperatures as were the bars in the first series, but these were annealed to give a maximum of spheroidal cementite. Tensile tests were also made on these. The final temperature of annealing was the same in both cases, i. e., 1500 degrees Fahr., but the bars of the first series were first heated to 1675 degrees Fahr. for three hours. The results of the tensile tests on these bars are shown in Table XIII. Comparing any bar in the first series, that is, any of specimens 401, 402, or 403, with any bar in the second series, that is, specimens 404, 405, and 406, it will at once be apparent that the bars in the first series have much lower reduction of area, elongation, and higher breaking loads than have the bars in the second series. These differences are due entirely to the annealing which followed the high temperature treatment. A comparison of the bars having the same anneal but different previous treatments, that is a comparison of bar 401 with bars 402 and 403, indicates that the previous high temperature treatment had no effect on the tensile properties of these bars. The same situation exists for bars 404, 405, and 406. As a result of these tests we are led to believe that heating into the range from 1700 to 2200 degrees Fahr. and following this by proper annealing have

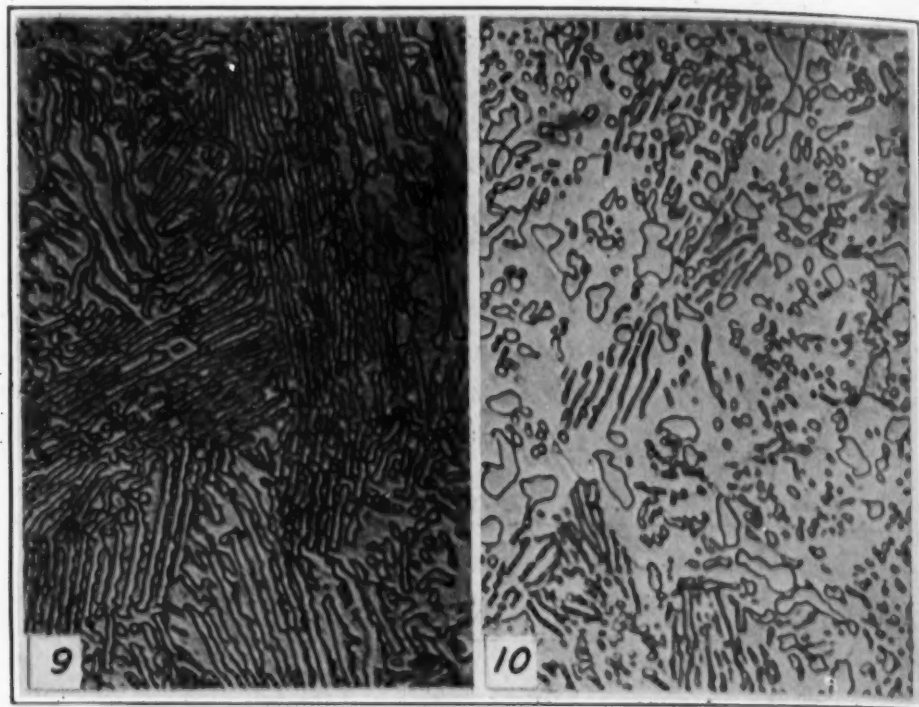


Fig. 9—Specimen 359-B, a Sample of S. A. E. 1085 Steel Heated to 2400 and Annealed at 1500 Degrees Fahr.

Fig. 10—Specimen 406-B, a Sample of S. A. E. 1085 Steel Heated to 1700 Degrees Fahr., and Annealed at 1500 Degrees Fahr.

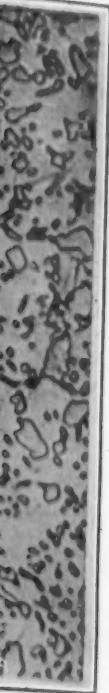
Specimens Etched with 4 Per Cent Nitric Acid in Alcohol: Magnifications  $\times 1250$ .

no measurable ill effects on the tensile properties of S.A.E. 1085 steel.

A number of the test specimens of this steel listed in Tables XI, XII, and XIII were sectioned and subjected to microscopic examination. Specimen 395B, which had been heated to 2400 degrees Fahr. before annealing, is shown in the Fig. 9. Here the pearlite is mainly lamellar. Fig. 10 shows the microstructure of specimen 406B to be mainly spheroidal. This sample was heated to 1700 degrees Fahr. and then annealed at 1500 degrees Fahr. and cooled rather slowly. When this spheroidal structure is produced, the steel is more ductile than when lamellar cementite is obtained. The formation of this type of pearlite is due to certain methods of annealing, and, consequently, the ductility of the steel in this case is due to the annealing rather than to the heating previous to annealing. There were no structural differences in any of the sections examined that appeared to be due to excessive heating.

It was thought that, although no ill effect might result from the





2400 and  
00 Degrees  
1250.

E. 1085

bles XI,  
exami-  
es Fahr.  
mainly  
B to be  
s Fahr.  
slowly.  
ductile  
of this  
conse-  
nealing  
ere no  
appeared  
om the

Table X  
Summary of Tensile, Hardness and McQuaid-Ehn Tests on S.A.E. 1085 Steel

Specimen	Heat Treatment Deg. F.	Anneal 40 Min. Deg. F.	Yield Point #/in. <sup>2</sup>	Tensile Strength #/in. <sup>2</sup>	Breaking Load #/in. <sup>2</sup>	% Elongation	% Reduction of Area	Sclero- scope Hardness	B Scale Rockwell Hardness	Brinell Hardness	Ehn Grain Size
310	A	2550	1500	111,350	111,350	5.75	6.2	28	93	228	10-11
	B	2550	1500	108,450	108,450	5.75	6.2	27	95	228	11
	C	2550	1500	107,850	107,850	6.5	7.0	27.5	94	228	..
	Av.	2550	1500	109,200	109,200	6.0	6.5	27.5	94	228	..
311	A	2450	1500	116,000	116,000	6.5	7.2	29	95	241	11
	B	2450	1500	114,800	114,800	5.75	7.0	28	94	228	11
	C	2450	1500	111,425	111,425	5.4	5.4	28.5	94.5	235	..
	Av.	2450	1500	114,075	114,075	5.67	6.53	28.5	94.5	235	..
312	A	2300	1500	114,300	112,125	10.0	12.6	29	95	228	10
	B	2300	1500	113,350	109,650	10.0	13.8	27	91	223	10-11
	C	2300	1500	120,500	119,000	10.0	12.0	28	93.0	226	..
	Av.	2300	1500	116,050	113,592	10.0	13.1	28	93.0	226	..
313	A	2200	1500	113,750	111,100	10.5	10.7	23	92	223	5-11
	B	2200	1500	116,250	113,100	10.0	13.3	26	92	223	4-11
	C	2200	1500	108,700	104,250	8.25	7.1	25	92	223	..
	Av.	2200	1500	112,900	109,483	10.25	12.0	25	92	223	..
314	A	None	1500	93,500	88,100	14.25	28.5	26	89	192	5-9
	B	None	1500	94,600	91,000	13.5	23.0	20	83	183	6-9
	C	None	1500	96,050	92,950	13.0	27.5	23	86	187	..
	Av.	None	1500	94,717	90,683	13.6	28.0	23	86	187	..

Chemical Analysis of the Steel:

Carbon	0.88 %
Manganese	0.34 %
Sulphur	0.018 %
Phosphorus	0.029 %

Table XI  
Summary of Tensile and Hardness Tests on S.A.E. 1085 Steel

Specimen	Heat Treatment Deg. F.	Annealed 40 Min. Deg. F.	Yield Point #/in. <sup>2</sup>	Tensile Strength #/in. <sup>2</sup>	Breaking Load #/in. <sup>2</sup>	% Elongation in 2 in.	% Reduction of Area	Sclero- scope Hardness	B Scale Rockwell Hardness	Brinell Hardness
359 A	2400	1500	46,450	103,650	103,450	15.5	19.2	25	91	205
B	2400	1500	45,154	107,879	104,375	15.0	17.6	25	93	215
C	2400	1500	45,800	106,765	104,010	15.2	18.4	25	92.0	210
Av.										
360 A	2385	1500	48,300	109,350	106,400	14.0	18.8	27	93	192
B	2385	1500	47,500	110,865	108,460	12.0	17.4	27	94	200
C	2385	1500	47,500	111,950	109,100	13.0	17.7	27	93.5	196
Av.										
361 A	2350	1500	46,500	111,450	108,050	13.0	16.6	29	95	217
B	2350	1500	46,100	112,900	109,700	12.5	16.3	28	95	215
C	2350	1500	46,300	114,650	111,350	13.0	17.0	28.5	95	216
Av.										
362 A	2320	1500	46,556	115,538	113,336	11.5	14.5	28	95	223
B	2320	1500	47,614	116,780	113,873	10.5	14.5	28	96	228
C	2320	1500	48,900	117,700	115,900	10.0	13.7	28	95.5	225
Av.										
363 A	2290	1500	47,690	116,673	114,370	10.7	14.2	29	96	223
B	2290	1500	49,000	117,650	115,350	11.0	14.8	29	97	223
C	2290	1500	48,700	118,650	116,150	12.0	14.8	29	96.5	223
Av.										
364 A	1900	1500	47,266	102,643	99,790	16.0	20.8	24	91	196
B	1900	1500	48,750	102,100	98,600	15.5	20.6	24	92	192
C	1900	1500	48,350	105,500	101,350	15.5	21.3	24	91.5	194
Av.										

## Chemical analysis of the Steel:

Carbon	0.86 %
Manganese	0.27 %
Sulphur	0.018 %
Phosphorus	0.029 %

excessive heating of annealed specimens, others in the hardened and tempered condition might be affected by previous excessive heating, and also that the hardness after quenching might be affected. To study this effect on the hardness, three bars of the S.A.E. 1085 steel were heated to 2400 degrees Fahr., three bars to 2300 degrees Fahr., and three bars were not heated. These nine bars were then all annealed at the same time at 1500 degrees Fahr., then reheated to 1500 degrees Fahr. in the same furnace at the same time and quenched in water. After quenching, hardness tests were made at the center of all the bars. The results of these tests are shown in Table XIV. These results would indicate that previous excessive heating to temperatures as high as 2400 degrees Fahr. has no effect on the hardness of steel if the steel is subsequently annealed and quenched.

The results of the tests made to determine the effect of excessive heating on S.A.E. 1085 steel which is subsequently hardened and tempered are given in Table XV. The data in this table shows that heating as high as 2400 degrees Fahr. previous to annealing, hardening, and tempering has little effect on the tensile properties of this steel. There is, however, a somewhat lower reduction of area and elongation in the bars heated to 2400 degrees Fahr. than was found for those heated to 2310 degrees Fahr.

In Table X it will be noted that the grain size as obtained in the McQuaid-Ehn test is finer in the S.A.E. 1085 steel previously heated to 2300 degrees Fahr. and above than in those heated below this temperature. Since this was noted on only a few samples it was decided to make further tests of this type. Accordingly, a number of McQuaid-Ehn tests were made on other samples of this steel and check tests were made on some of the samples already reported.

In making these tests samples  $\frac{5}{8}$  inches long were cut from the tensile bars of the S.A.E. 1085 steel after the tensile tests had been made. The samples were cut from the  $\frac{3}{4}$ -inch round part of the tensile bars of the S.A.E. 1085 steel after the tensile tests had heated in carburizing material for six hours at 1725 degrees Fahr., and then the pot was removed from the furnace, and cooled in the air. After cooling, the samples were removed from the pot and sectioned, polished, etched, and examined under the microscope. The grain sizes were compared with those in the chart prepared by J. D. Gat<sup>5</sup> and numbered according to his system. The results of these

<sup>5</sup>J. D. Gat, "Normality of Steel," *TRANSACTIONS, American Society for Steel Treating*, Vol. 12, 1927, p. 376.

Table XII  
Summary of Tensile and Hardness Tests on S.A.E. 1095 Steel

Specimen	Heat Treatment Deg. F.	Anneal 40 Min. Deg. F.	Yield Point #/in. <sup>2</sup>	Tensile Strength #/in. <sup>2</sup>	Breaking Load #/in. <sup>2</sup>	% Elongation in 2 in.	% Reduction of Area	Sclero- scope Hardness	B Scale Rockwell Hardness	Brinell Hardness
400 A	2300	1500	46,200	102,300	97,500	16.5	23.4	29.0	88	196
B	2300	1500	46,500	101,950	97,600	16.5	24.3	29	91	196
Av.	2300	1500	46,350	102,125	97,550	16.5	23.8	29	89.5	196
399 A	2200	1500	49,500	104,250	100,400	15.0	19.4	28	87	207
B	2200	1500	48,300	103,250	98,900	16.0	22.2	27	87	192
C	2200	1500	47,250	102,100	97,400	17.0	22.7	27.5	87	200
Av.	2200	1500	48,350	103,200	98,900	16.0	21.4	28	87	200
398 A	2100	1500	49,150	106,850	101,850	14.0	19.7	28	89	207
B	2100	1500	47,300	106,250	102,500	14.0	19.1	28	90	212
Av.	2100	1500	48,225	106,550	102,175	14.0	19.5	28	89.5	210
397 A	2010	1500	46,700	103,500	99,150	17.0	25.0	28	89	207
B	2010	1500	47,374	106,465	101,320	16.0	21.5	28	89	207
C	2010	1500	47,040	104,980	100,230	16.5	23.2	28	89	207
Av.	2010	1500	46,700	104,820	99,890	16.5	23.2	28	89	207
396 A	1895	1500	45,250	104,400	100,250	15.75	21.0	26	89	205
B	1895	1500	47,500	103,650	99,600	15.5	21.6	28	89	205
C	1895	1500	49,500	104,000	99,150	15.85	22.7	27	89	203
Av.	1895	1500	47,420	104,020	99,670	15.7	21.8	27	89	203
395 A	1725	1500	49,100	100,350	94,700	20.0	28.4	27	89	196
B	1725	1500	47,750	102,450	97,200	18.0	24.4	27	89	192
C	1725	1500	46,944	102,104	96,590	16.5	24.3	27	89	194
Av.	1725	1500	47,930	101,635	96,180	18.2	25.7	27	89	194

## Chemical Analysis of the Steel:

Carbon	0.99 %
Manganese	0.31 %
Sulphur	0.035 %
Phosphorus	0.020 %



Table XIII  
Summary of Tensile and Hardness Tests on S.A.E. 1085 Steel

Specimen	Heat Treatment Deg. F.	Annealed 40 Min. Deg. F.	Yield Point #/in. <sup>2</sup>	Tensile Strength #/in. <sup>2</sup>	Breaking Load #/in. <sup>2</sup>	% Elongation in 2 in.	% Reduction of Area	Sclero- scope Hardness	B Scale Rockwell Hardness	Brinell Hardness
401 A	2090	1500, 1675	41,850	103,510	101,230	14.5	17.5	26	90	207
B	2090	1500	40,970	103,510	101,230	14.5	17.5	27	89	207
C	2090	"	43,450	104,500	100,950	15.0	17.5	26.5	89.5	207
Av.	2090	"	42,090	104,005	101,090	14.8	17.5	28	90	209
402 A	2000	"	40,000	105,800	101,600	14.0	17.9	28	90	207
B	2000	"	40,640	103,380	101,600	14.0	16.0	28	90	207
C	2000	"	40,830	105,210	103,550	14.0	17.0	28	90	208
Av.	2000	"	40,490	104,800	102,575	14.0	17.0	29	89	207
403 A	1700	"	41,250	104,350	102,050	14.5	17.3	27	89	196
B	1750	"	41,500	103,200	100,650	14.5	18.7	27	89	196
C	1750	"	43,200	105,650	103,350	14.0	16.9	28	89	201
Av.	1750	"	41,980	104,400	102,020	14.3	17.6	27	85	187
404 A	2100	1500	53,200	94,000	89,100	22.0	30.8	26	86	187
B	2100	1500	56,200	93,000	85,650	24.5	37.3	26	85	187
C	2100	1500	51,950	92,950	84,700	24.0	37.6	26.5	85.5	187
Av.	2100	1500	53,780	93,320	86,480	23.5	35.2	25	86	187
405 A	2000	1500	51,850	93,500	88,300	23.0	32.7	26	85	179
B	2000	1500	52,835	92,830	86,170	24.5	34.9	25	85	179
C	2000	1500	54,800	93,100	87,250	23.7	34.4	25.5	85.5	183
Av.	2000	1500	53,160	93,140	87,240	23.7	34.0	27	85	183
406 A	1700	1500	50,750	93,100	85,000	23.7	37.6	25	86	183
B	1700	1500	51,540	92,350	84,500	24.5	37.4	25	86	183
C	1700	1500	55,100	93,500	87,550	24.2	33.4	25	85.5	183
Av.	1700	1500	52,460	92,980	85,680	24.1	36.1	25	85.5	183

Chemical Analysis of the Steel:

Carbon	0.90 %
Manganese	0.31 %
Sulphur	0.035 %
Phosphorus	0.020 %

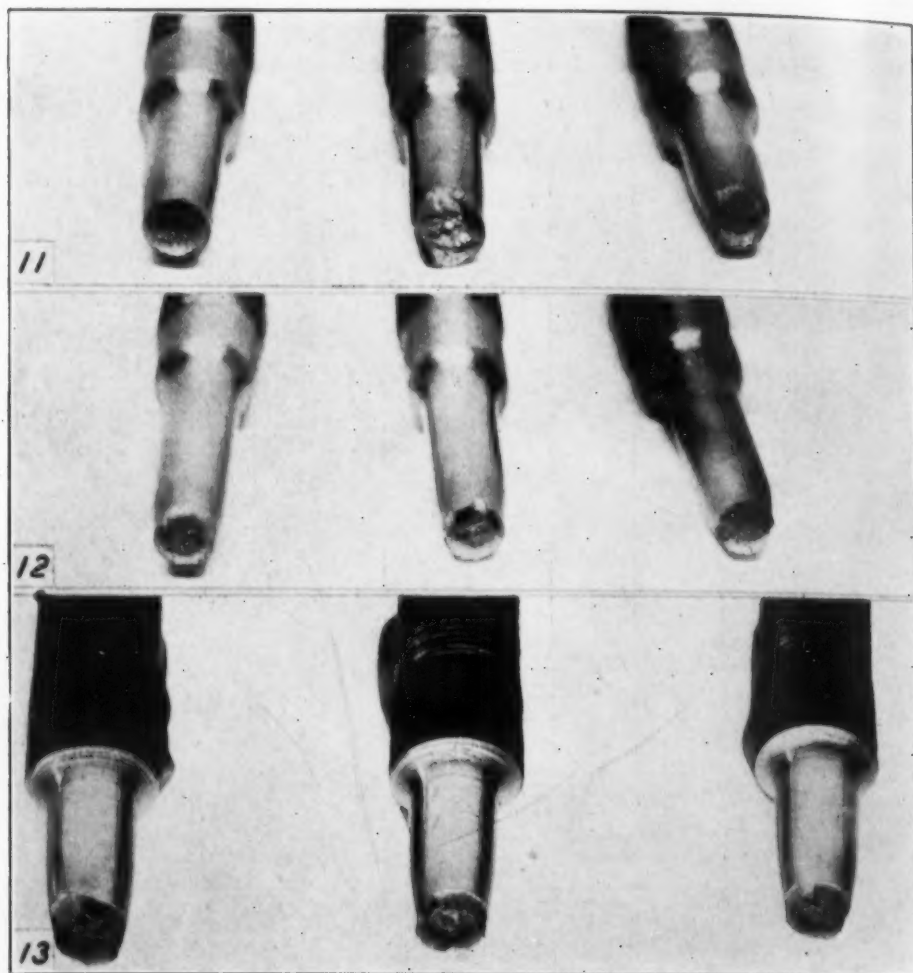


Fig. 11—Specimens 326 of S. A. E. 3145 Steel Heated to 2535 Degrees Fahr., and Annealed at 1550 Degrees Fahr. Note that there is Some Coarseness in the Grain of This Fracture.

Fig. 12—Specimens 309 of S. A. E. 3145 Steel not Heated Previous to Annealing at 1550 Degrees Fahr. Note the Fine Grained Cup Shaped Fracture.

Fig. 13—Specimens 378 of S. A. E. 3145 Steel Heated to 2540 Degrees Fahr., Annealed at 1550 Degrees Fahr., Heated to 1450 Degrees Fahr., and Quenched into Oil and Reheated to 1050 Degrees Fahr.

tests are presented in Table XVI. It will be observed by reference to this table that the temperature of previous heating does not seem to be related to the McQuaid-Ehn grain size, and therefore, the observation in Table X probably was the result of some unaccountable circumstance.

*Tests on S.A.E. 3145 Steel.* Table XVII and XVIII, contain the results of two series of tests on S.A.E. 3145 steel. The data given seems to show a gradual change in ductility in the test bars of

this steel beginning at 2450 degrees Fahr., the ductility being less as the temperature of overheating is increased. It will be observed from these tables that the lowering of the per cent elongation and per cent reduction of area takes place over a range of temperature. The S.A.E. 3145 steel is the only one of the group of steels studied which behaves in this manner. Together with the decrease in ductility there appears to be an increase in the breaking loads. This phenomenon is also accompanied by a difference in the appearance of the fracture. The appearance of the fracture of the three bars heated to 2535 degrees Fahr. and subsequently annealed are displayed in Fig 11. These fractures should be compared with those of the three bars not heated previous to annealing which are shown in Fig. 12. The McQuaid-Ehn tests showed no difference of grain size as a result of excessive heating of this steel.

A series of tests were made to determine the behavior of the S.A.E. 3145 steel in the hardened and tempered condition rather than in the annealed condition, when the steel was previously overheated. In this series of tests three bars of this steel were heated to 2540 degrees Fahr., three to 2500 degrees Fahr., three to 2475 degrees Fahr., three to 2450 degrees Fahr., and three were not heated. All of these bars were then annealed together at 1550 degrees Fahr., then reheated to 1450 degrees Fahr. and quenched in oil, and after this heated to 1050 degrees Fahr. They were then machined into 0.505-inch diameter tensile bars and tested. A summary of these tests is presented in Table XIX. The same characteristics are shown here as were obtained in the two previous series of tests in which the steel was annealed after being heated to the relatively high temperatures. This is shown in Table XIX by the decrease in elongation and reduction of area, and by the increase in the breaking load beginning with the bars heated to 2450 degrees Fahr. This change in physical properties indicates a loss in the ductility of the metal. The appearance of the fracture also showed a change corresponding to the change in the temperature of excessive heating. Beginning with bars heated to a temperature of 2450 degrees Fahr. there is a change in the appearance of the fracture from star-shape which appear on the bars heated below this temperature to cup-shape which appears on the bars heated to this temperature or above. It is interesting to note that this steel in the annealed condition shows cup-shaped fractures when not previously excessively heated, whereas, in the hardened

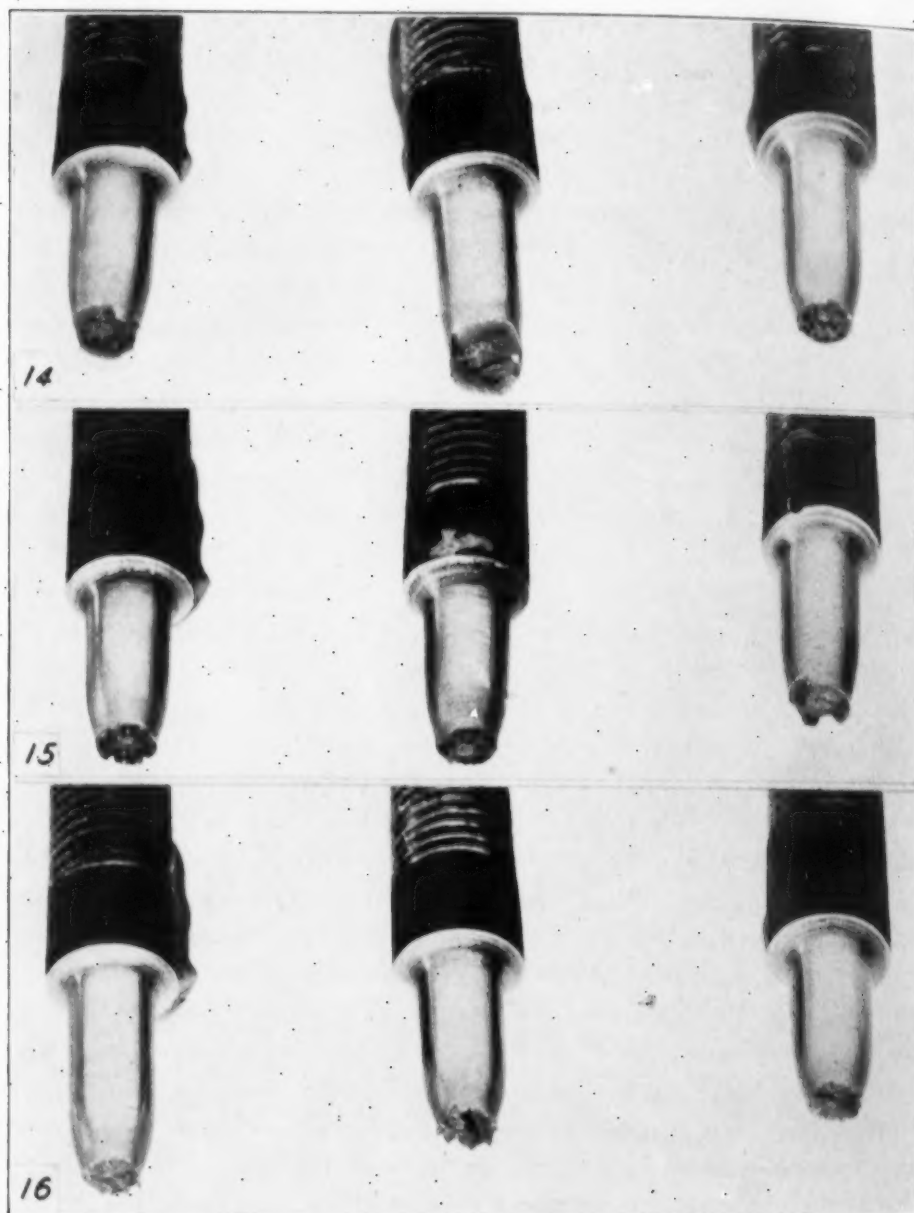


Fig. 14.—Specimens 384 of S. A. E. 3145 Steel Heated to 2500 Degrees Fahr., Annealed at 1550 Degrees Fahr., Heated to 1450 Degrees Fahr., and Quenched Into Oil and Reheated to 1050 Degrees Fahr.

Fig. 15.—Specimens 385 of S. A. E. 3145 Steel Heated to 2450 Degrees Fahr., Annealed at 1550 Degrees Fahr., Heated to 1450 Degrees Fahr., and Quenched in Oil and Reheated to 1050 Degrees Fahr.

Fig. 16.—Specimens 379 of S. A. E. 3145 Steel not Heated Previous to Annealing at 1550 Degrees Fahr., then Heated to 1450 Degrees Fahr., and Quenched in Oil and Reheated to 1050 Degrees Fahr.

and tempered condition the fracture is star-shaped unless excessively heated.



**Table XIV**  
**Summary of Hardness Readings Made on Water-Quenched S.A.E. 1085 Steel**

Specimen	Heated (Sand air cool) Deg. F.	Annealed (Furnace cooled) Deg. F.	Water quenched from 1500° F.		
			Brinell Hardness	Rockwell "C"	Scleroscope Hardness
386 A	2400	1500	652	67	75
B	2400	1500	627	67	67
Av.	2400	1500	639	67	71
387 A	2300	1500	642	67	69
B	2300	1500	642	67	69
Av.	2300	1500	642	67	69
388 A	No heat	1500	642	68	69
B	No heat	1500	652	68	70
Av.	No heat	1500	647	68	69.5

Chemical Analysis of the Steel:

Carbon .....	0.81 %
Manganese .....	0.28 %
Sulphur .....	0.018%
Phosphorus .....	0.029%

Figs. 13 through 16 show the fractures of these test bars in the hardened and tempered condition. It will be observed by examining these photographs that specimen 378A, B, and C, shown in Fig. 13, has cup-shaped fractures. These specimens were heated to 2540 degrees Fahr. previous to hardening and tempering. Specimens 379A, B, and C, Fig. 16, which were not heated previous to hardening and tempering, show all fractures to be star-shaped. Specimens 384, Fig. 14, heated to 2500 degrees Fahr. and Specimen 385, Fig. 15, heated to 2450 degrees Fahr. show variations of both types of fractures. There thus appears to be a relationship between the temperature of excessive heating, the ductility, and the shape of the fractures.

Tests were run in order to determine whether heating to these high temperatures would affect the impact resistance values as measured by the Izod impact testing machine. In these tests the samples were all annealed at 1550 degrees Fahr. after the high-temperature treatment as shown in Table XX. An examination of this table does not show any ill effect on resistance to impact to result from the excessive heating.

A series of tests were made to determine whether there might be a change in the electrical conductivity of the test bars that had been excessively heated. If such a change were found this method would be an easy one for determining whether or not steel was excessively heated. Accordingly, three test bars were heated to 2550 degrees Fahr., three to 2500 degrees Fahr., three to 2450 degrees Fahr., and three were not heated. These bars were all annealed together at

Table XV  
Summary of Tensile and Hardness Tests on S.A.E. 1085 Steel, Heated, Quenched, and Tempered

Specimen	Heated Deg. F.	Normalized 40 Min. Deg. F.	Quenched (oil) Deg. F.	Tempered 1 hr. Deg. F.	Yield Point Lbs./in. <sup>2</sup>	Tensile Strength #/in. <sup>2</sup>	Breaking Load #/in. <sup>2</sup>	% Elongation in 2 in.	% Reduction of Area	Hardness (after Tempering) Brinell scale	Rockwell "C" scale	Selec- roscope
391 A	2400	1500	1425	1100	86,750	135,000	121,550	14.5	33.8	272	27	35
B	2400	1500	1425	1100	83,150	135,750	123,450	15.0	33.1	269	27	36
C	2400	1500	1425	1100	87,700	136,950	125,000	15.0	32.8	270	27	35.5
Av.	2400	1500	1425	1100	85,870	135,900	123,660	14.8	33.2	270	27	37
392 A	2310	1500	1425	1100	92,700	133,950	110,150	17.5	45.2	269	27	37
B	2310	1500	1425	1100	93,950	135,700	116,750	16.0	39.1	269	27	37
C	2310	1500	1425	1100	82,300	137,850	120,850	17.5	38.9	269	27	37
Av.	2310	1500	1425	1100	89,650	135,835	115,920	17.0	41.1	269	27	37
393 A	2100	1500	1425	1100	89,550	133,400	110,000	18.0	45.5	255	26	35
B	2100	1500	1425	1100	91,600	134,850	106,400	18.5	50.0	262	25	35
C	2100	1500	1425	1100	93,850	138,850	112,550	18.5	46.0	258	25.5	35
Av.	2100	1500	1425	1100	91,670	135,700	109,650	18.3	47.2	258	25.5	35
394 A	1900	1500	1425	1100	82,000	135,050	115,550	17.5	41.9	269	26	35
B	1900	1500	1425	1100	83,600	137,100	113,600	19.0	44.6	269	28	34
C	1900	1500	1425	1100	83,850	142,850	122,700	18.0	36.6	269	27	34.5
Av.	1900	1500	1425	1100	83,150	138,330	117,280	18.2	41.0	269	27	34.5

## Chemical Analysis of the Steel:

Carbon	0.86 %
Manganese	0.26 %
Sulphur	0.18 %
Phosphorus	0.029 %

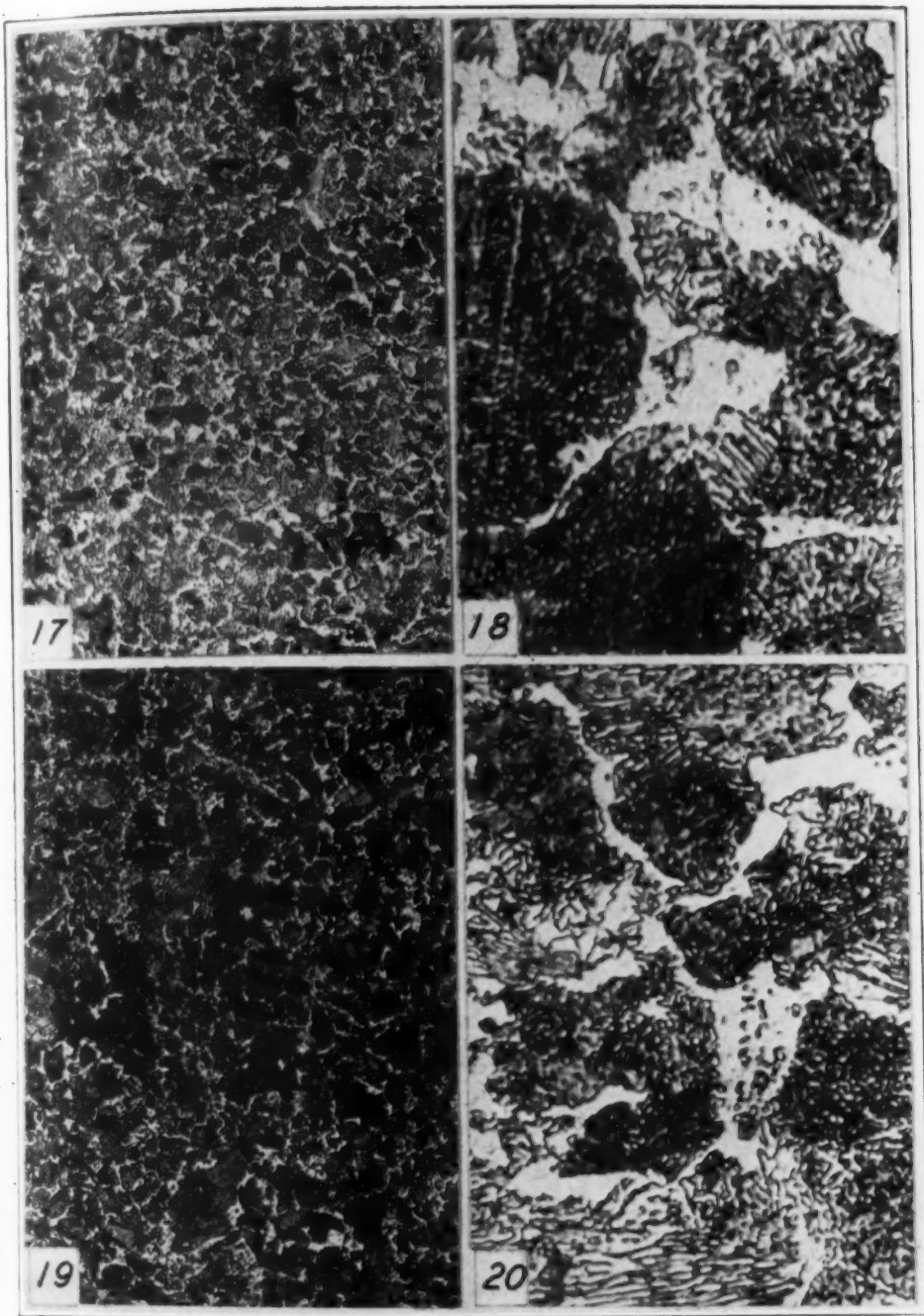


Fig. 17—Specimen 326A, a sample of S. A. E. 3145 Steel Heated to 2535 Degrees Fahr., and Annealed at 1550 Degrees Fahr. Magnification  $\times 150$ .

Fig. 18—Specimen 326A, a Sample of S. A. E. 3145 Steel Heated to 2535 Degrees Fahr., and Annealed at 1550 Degrees Fahr. Magnification  $\times 1250$ .

Fig. 19—Specimen 325A, a sample of S. A. E. 3145 Steel Heated to 2500 Degrees Fahr., for 2 Hours and Annealed at 1550 Degrees Fahr. Magnification  $\times 1250$ .

Fig. 20—Specimen 325A, a Sample of S. A. E. 3145 Steel Heated to 2500 Degrees Fahr., for 2 Hours and Annealed at 1550 Degrees Fahr. Magnification  $\times 1250$ .

Etched with 4 Per Cent Nitric Acid in Alcohol.

1550 degrees Fahr. The bars were then machined to  $\frac{3}{8}$  inches square by  $7\frac{1}{2}$  inches long and measured for electrical resistance by means of an electrical set-up which used the double Kelvin bridge. Table XXI shows the results obtained in this series of tests. As may be observed from this table, there is a small difference between the resistivity of the bars not heated and those heated to 2450 degrees

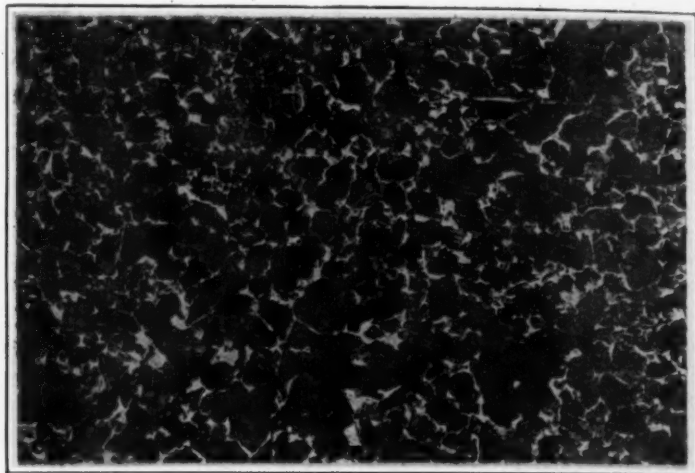


Fig. 21—Specimen 341B, a sample of S. A. E. 3145 Steel Heated to 2450 Degrees Fahr., and Annealed at 1550 Degrees Fahr. Magnification  $\times 150$ . Etched with 4 Per Cent Nitric Acid in Alcohol.

Fahr. or higher. The difference, however, is so small that it would not be practical to attempt to use this method to indicate excessive heating.

A metallographic study was made of a number of the S.A.E. 3145 steels in an effort to find an explanation of their behavior when heated to high temperatures.

Specimen 305A, which had been heated to 2600 degrees Fahr. before annealing, was found to be burned. Specimen 327A, which had been heated to 2550-2580 degrees Fahr. and annealed, gave no evidence of being burned, but showed persistent grain boundaries which the anneal had failed to remove. Specimen 306A, which had been heated to 2500 degrees Fahr. and annealed, also showed some persistent grain boundaries.

Specimen 326A, which was heated to 2535 degrees Fahr. and annealed, shows some evidence of persistent grain boundaries although they are thin and not easily distinguished. Fig. 17 shows a section of this structure with some traces of gamma-iron grain



1929

Table XVI  
Summary of Results Obtained with the McQuaid-Ehn Test on S.A.E. 1085 Steel

Specimen	Bar No.	Heat Treatment Deg. F.	Anneal 40 min. Deg. F.	McQuaid- Ehn Grain Size
320A	B1	2400	1500	11
320B	B1	2400	1500	11
320C	B1	2400	1500	11
321A	B1	2375	1500	11
321B	B1	2375	1500	11
321C	B1	2375	1500	11
322A	B1	2350	1500	10
322B	B1	2350	1500	10-11
322C	B1	2350	1500	10
323A	B1	2325	1500	10
323B	B1	2325	1500	10
323C	B1	2325	1500	9-10
324A	B1	2300-2 hrs.	1500	11
324B	B1	2300-2 hrs.	1500	10
324C	B1	2300-2 hrs.	1500	10-11
360B	B4	2385	1500	10-11
361B	B4	2350	1500	10
362B	B4	2320	1500	10
363B	B4	2290	1500	10-11
364A	B4	1900	1500	10
364C	B4	1900	1500	6
312B	B5	2300	1500	10
313A	B5	2200	1500	10
313B	B5	2200	1500	10
313C	B5	2200	1500	9-10
314C	B5	Not heated	1500	9-10
400B	C1	2300	1500	5-8
399B	C1	2200	1500	6-7
398B	C1	2100	1500	6-7
397B	C1	2010	1500	6
396B	C1	1895	1500	6-7
395B	C1	1725	1500	7

boundaries. Fig. 18 shows this structure at 1250 diameters. Fig. 19 which shows the structure of specimen 325A, also shows some traces of persistent grain boundaries. Specimen 325A, had been heated to 2500 degrees Fahr. for two hours and annealed. Fig. 20 shows the structure of specimen 325A at 1250 diameters.

Specimen 341B, which was heated to 2450 degrees Fahr. and annealed, is shown in Fig. 21. No persistent grain boundaries are visible in this microstructure.

Specimens 384B and 385A, which were heated to 2500 and 2450 degrees Fahr., respectively, then annealed, reheated, and quenched from 1450 degrees Fahr. into oil, and tempered at 1050 degrees Fahr., were examined and compared with specimen 379A, which was not heated to a high temperature, but was given the same subsequent treatment with the other samples. These specimens all appeared alike under the microscope both at 150 and at 1250 diameters magnification.

It appears from examining these microstructures of the various steels that excessively heating the steel changes the microstructure

Table XVII  
Summary of Tensile, Hardness and McQuaid-Ehn Tests on S.A.E. 3145 Steel

Specimen	Heat Treatment Deg. F.	Anneal 40 Min. Deg. F.	Tensile Strength #/in. <sup>2</sup>	Breaking Load #/in. <sup>2</sup>	Elongation % in 2 in.	% Reduction of Area	Sclero- scope Hardness	B Scale Rockwell Hardness	Brinell Hardness	McQuaid- Ehn Grain Size
305 A	2600	1550	100,000	99,200	15.0	17.7	23	91	196	N 10
	2600	1550	99,350	98,900	15.0	17.7	23	91	196	N 10
	2600	1550	99,050	97,675	14.0	17.4	24	90	196	
	2600	1550	99,533	98,592	14.67	17.6	23	91	196	
306 A	2500	1550	100,000	90,700	21.5	34.7	24	91	196	N 9-10
	2500	1550	100,200	89,400	21.5	35.4	23	91	195	N 10
	2500	1550	99,800	88,300	22.0	36.3	24	91	192	
	2500	1550	100,000	89,467	21.7	35.5	24	91	195	
307 A	2400	1550	99,250	81,675	23.8	48.1	23	90	196	N 10
	2400	1550	100,150	80,250	23.6	48.4	23	91	196	N 10
	2400	1550	97,850	80,950	23.5	46.9	23	90	196	
	2400	1550	99,083	80,958	23.6	47.8	23	90	196	
308 A	2300	1550	97,300	81,075	25.5	46.0	23	90	192	N 10
	2300	1550	97,350	81,950	24.3	43.7	23	90	187	N 9-10
	2300	1550	96,500	82,300	23.8	45.8	23	90	187	
	2300	1550	97,050	81,775	24.5	45.2	23	90	189	
309 A	None	1550	96,450	81,925	24.0	43.4	22	90	187	N 10
	None	1550	96,500	77,450	24.0	43.7	22	90	187	N 9
	None	1550	96,200	80,450	24.0	43.7	23	90	183	
	None	1550	96,383	80,000	24.0	43.6	22	90	186	

## Chemical Analysis of the Steel:

Carbon	0.45 %
Manganese	0.61 %
Sulphur	0.019 %
Phosphorus	0.018 %
Silicon	0.252 %
Nickel	1.27 %
Chromium	0.53 %

and that these changes seem to accompany the changes in physical properties. The hardened and tempered specimens were the exception to this since they did not show a change in microstructure although they had varying physical properties related to the tempera-

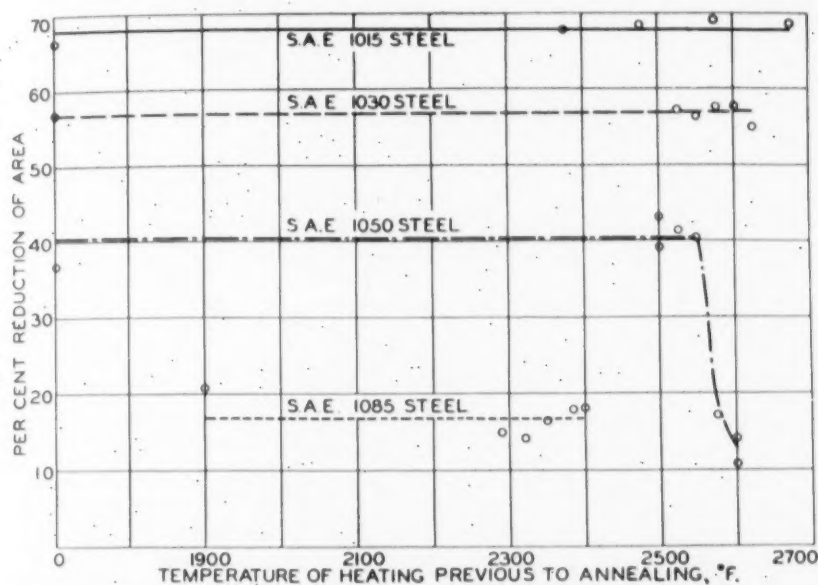


Fig. 22—The Relation Between the Per Cent Reduction of Area and the Temperature of Excessive Heating Previous to Annealing for the Four Plain Carbon Steels Studied.

ture of overheating. It is, however, quite common to have structural differences masked in the hardened condition—even pronounced segregation is often masked by a hardening treatment.

It was felt that perhaps some other etching reagents might be used to bring out the persistent grain boundaries or other structural differences. Accordingly, samples of the S.A.E. 3145 steel which had shown the persistent grain boundaries and some which had not been heated and which, consequently, showed none of these grain boundaries, were etched with about 15 different etching reagents.

Although various microstructures were brought out by these etching reagents, none of them showed any difference between the excessively heated samples and those not heated, other than what had already been brought out by the nitric acid etch.

From these tests it appears that although there have been certain differences in the microstructure of excessively heated steel, the microstructure cannot be relied on to indicate definitely whether or not a steel has been excessively heated, and that the safest test is the tensile test.

Table XVIII  
Summary of Tensile, Hardness and McQuaid-Ehn Tests on S.A.E. 3145 Steel

Specimen	Heat Treatment Deg. F.	Annealed 40 Min. Deg. F.	Yield Point #/in. <sup>2</sup>	Tensile Strength #/in. <sup>2</sup>	Breaking Load #/in. <sup>2</sup>	% Elongation	% Reduction of Area	Sclero- scope Hardness	B Scale Rockwell Hardness	Brinell Hardness	McQuaid- Ehn Grain Size
329 A	2600	1550	61,750	106,900	97,000	19.5	36.6	20	90	192	11-12
B	2600	1550	62,750	107,200	97,400	19.5	32.8	21	91	196	..
C	2600	1550	63,000	106,850	96,000	21.5	36.0	..	..	..	..
Av.	2600	1550	62,500	106,950	96,800	20.2	35.1	20.5	90.5	194	..
328 A	2590	1550	62,900	107,600	97,500	18.5	34.4	23	89	196	11
B	2590	1550	61,850	107,150	95,200	20.0	35.7	24	88	196	..
C	2590	1550	64,250	106,850	97,750	19.0	35.4	..	..	..	..
Av.	2590	1550	62,670	107,200	96,820	19.2	35.2	23.5	88.5	196	..
327 A	2575	1550	62,700	105,150	100,350	18.5	25.1	24	89	187	10-11
B	2575	1550	61,625	105,300	99,500	19.5	24.8	25	91	..	..
C	2575	1550	63,250	106,750	96,750	20.0	33.8	..	..	..	..
Av.	2575	1550	62,525	105,730	98,865	19.3	27.9	24.5	90	187	..
339 A	2550	1550	62,450	106,350	98,600	19.0	31.8	22	91	187	10
B	2550	1550	62,400	105,900	97,250	18.5	33.1	21	91	192	..
C	2550	1550	60,950	104,750	95,950	19.5	36.0	..	..	..	..
Av.	2550	1550	61,930	105,670	97,270	19.0	33.6	21.5	91	190	..
326 A	2535	1550	61,150	101,700	86,500	23.5	44.6	23	87	196	11
B	2535	1550	65,500	104,250	96,500	20.5	33.1	23	89	192	..
C	2535	1550	61,500	103,350	90,000	23.5	44.0	..	..	..	..
Av.	2535	1550	62,715	103,165	91,000	22.5	40.6	23	88	194	..
325 A	2500	1550	60,000	103,950	90,950	23.0	39.1	..	..	..	..
B	2 hrs.	1550	62,700	104,400	92,000	22.5	34.3	..	..	..	..
C	2 hrs.	1550	63,900	104,200	90,300	23.5	39.5	22	87	192	10-11
Av.	2 hrs.	1550	62,200	104,185	91,080	23.0	37.6	22	87	192	..



Table XVIII (Continued)  
Summary of Tensile, Hardness and McQuaid-Ehn Tests on S.A.E. 3145 Steel

Specimen	Heat Treatment Deg. F.	Annealed 40 Min. Deg. F.	Yield Point #/in. <sup>2</sup>	Tensile Strength #/in. <sup>2</sup>	Breaking Load #/in. <sup>2</sup>	% Elongation	% Reduction of Area	Sclero- scope Hardness	B Scale Rockwell Hardness	Brinell Hardness	McQuaid-Ehn Grain Size
340 A	2475	1550	61,150	105,500	87,650	24.5	46.0	21	90	189	10
B	2475	1550	64,650	105,500	88,500	24.5	44.0	20	91	191	..
C	2475	1550	63,000	105,650	90,750	22.0	43.1	..	..	..	..
Av.	2475	1550	62,900	105,550	88,970	23.6	44.4	20.5	90.5	190	..
341 A	2450	1550	62,750	105,750	86,050	24.5	48.1	21	86	192	..
B	2450	1550	64,600	106,300	85,900	23.0	50.0	21	88	196	..
C	2450	1550	64,055	106,575	88,000	21.5	49.0	..	..	..	..
Av.	2450	1550	63,800	106,210	86,650	23.0	49.0	21	87	194	..
357 A	2425	1550	62,750	107,457	88,712	22.5	48.4	25	92	196	..
B	2425	1550	62,350	106,500	88,100	23.0	48.4	23	92	196	..
C	2425	1550	63,400	106,200	87,650	22.0	48.9	..	..	..	..
Av.	2425	1550	62,835	106,719	88,154	22.5	48.6	24	92	196	..
358 A	2400	1550	62,500	106,050	86,850	23.0	48.4	23	93	198	..
B	2400	1550	63,365	106,366	90,650	22.0	45.4	23	92	201	..
C	2400	1550	64,000	106,850	85,200	23.0	51.7	..	..	..	..
Av.	2400	1550	63,290	106,420	87,570	22.7	48.5	23	92.5	200	..

Chemical Analysis of the Steel:

Carbon	0.44 %
Manganese	0.64 %
Sulphur	0.018 %
Phosphorus	0.011 %
Nickel	1.20 %
Chromium	0.74 %

Table XIX  
Summary of Tensile and Hardness Tests on S.A.E. 3145 Steel, Heated, Quenched, and Tempered

Specimen	Heated Deg. F.	Normalized		Tempered 1 hr. Deg. F.	Yield Point Lbs./in. <sup>2</sup>	Tensile Strength Lbs./in. <sup>2</sup>	Breaking Load Lbs./in. <sup>2</sup>	% Elongation in 2 in.	% Reduction of Area	Hardness (after Temper)	
		40 Min. Deg. F.	Quenched (oil) Deg. F.							Brinell	Rockwell "C" scale
378 A B C Av.	2540	1550	1450	1050	136,190	147,750	115,350	17.0	47.2	286	32
	10 min.	1550	1450	1050	138,750	147,135	120,940	15.5	46.1	281	33
	10 min.	1550	1450	1050	137,470	147,900	122,850	15.0	43.7	...	...
	10 min.	1550	1450	1050	136,500	147,595	119,710	15.8	45.7	283.5	32.5
384 A B C Av.	2500	1550	1450	1050	133,500	141,850	109,600	17.5	49.8	279	32
	2500	1550	1450	1050	129,000	141,550	114,550	12.5	34.1	255	35
	2500	1550	1450	1050	133,000	140,540	107,810	17.0	50.5	282	32
	2500	1550	1450	1050	133,000	141,310	110,650	15.7	44.8	272	32
390 A B C Av.	2475	1575	1450	1050	129,810	139,930	105,060	19.0	51.0	282	29
	2475	1575	1450	1050	129,460	140,430	101,830	19.0	54.2	286	30
	2475	1575	1450	1050	128,350	139,600	107,450	19.7	50.2	...	...
	2475	1575	1450	1050	129,205	139,985	104,780	19.3	51.8	284	29.5
385 A B C Av.	2450	1550	1450	1050	130,650	140,850	98,850	19.0	57.5	302	32
	2450	1550	1450	1050	134,350	141,550	100,700	19.0	56.0	298	34
	2450	1550	1450	1050	137,350	143,200	101,375	18.0	56.0	...	...
	2450	1550	1450	1050	134,120	141,870	100,310	18.7	56.5	300	33
379 A B C Av.	No heat	1550	1450	1050	139,900	148,560	96,770	19.0	60.8	286	32
	No heat	1550	1450	1050	143,700	151,900	99,250	...	58.3	292	33
	No heat	1550	1450	1050	149,500	158,500	103,750	...	60.1	...	...
	No heat	1550	1450	1050	144,370	152,990	99,920	19.0	59.7	289	32.5

## Chemical Analysis of the Steel:

Carbon	0.44 %
Manganese	0.64 %
Chromium	0.74 %
Nickel	1.20 %
Sulphur	0.018 %
Phosphorus	0.011 %

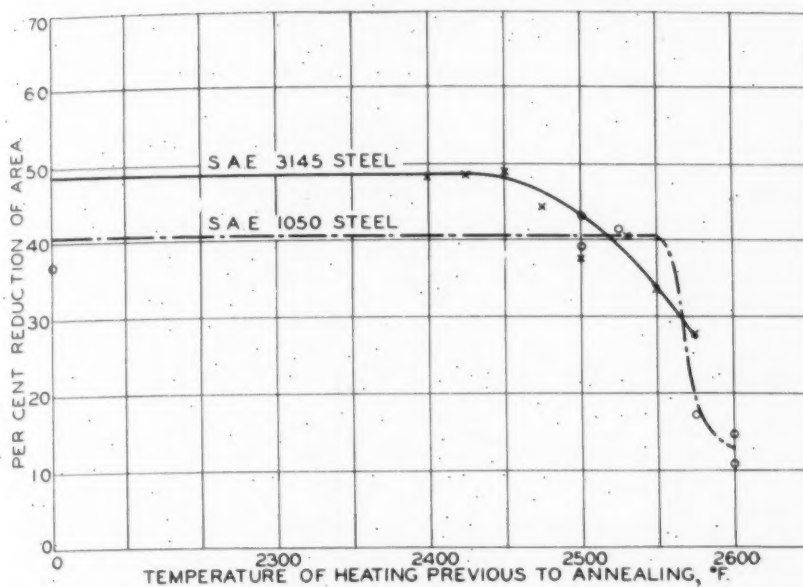


Fig. 23—The Relation Between the Per Cent Reduction of Area and the Temperature of Excessive Heating Previous to Annealing for S. A. E. 3145 Steel and S. A. E. 1050 Steel.

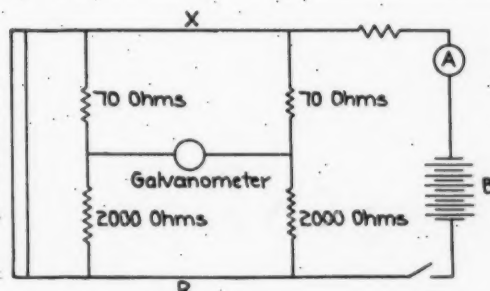
Table XXI  
Summary of Resistivity Tests of S.A.E. 3140 Steel

Specimen	Temperature to which steel was heated Deg. F.	Time in Furnace	Time at Temperature	Annealing Temperature Deg. F.	Temperature at which Resistivity was measured Deg. F.	Resistivity in microhms per square cm./cm.
410 A	2550	30 min.	10 min.	1550	72	29.81
B	2550	30 min.	10 min.	1550	72	29.79
C	2550	30 min.	10 min.	1550	72	29.70
Av.	2550	30 min.	10 min.	1550	72	29.77
411 A	2500	27 min.	10 min.	1550	72	29.76
B	2500	27 min.	10 min.	1550	72	29.74
C	2500	27 min.	10 min.	1550	72	29.67
Av.	2500	27 min.	10 min.	1550	72	29.72
412 A	2450	28 min.	10 min.	1550	72	29.74
B	2450	28 min.	10 min.	1550	72	29.74
C	2450	28 min.	10 min.	1550	72	29.72
Av.	2450	28 min.	10 min.	1550	72	29.73
413 A	Not heated	.....	.....	1550	72	29.54
B	Not heated	.....	.....	1550	72	29.46
C	Not heated	.....	.....	1550	72	29.39
Av.	Not heated	.....	.....	1550	72	29.46

Chemical Analysis of the Steel:

Carbon	0.38 %
Manganese	0.53 %
Chromium	0.61 %
Nickel	1.27 %
Sulphur	0.021 %
Phosphorus	0.028 %

R = Resistance in Ohms of Slide Wire.  
X = Resistance in Ohms of Specimen.



**Table XX**  
**Summary of Izod Impact Values for S.A.E. 3145 Steel**

Specimen	Temperature to which steel was heated Deg. F.	Time in Furnace (minutes)	Time at Heat (minutes)	Subsequent Heat Treatment	Impact Values (Ft. Lbs.)
365 A1	2605	26	10	Annealed at	28.5
A2	2605	26	10	1550°F.	23.5
A3	2605	26	10	1550°F.	30.0
Av.	2605	26	10	1550°F.	27.3
366 A1	2550	23	10	1550°F.	23.0
A2	2550	23	10	1550°F.	20.5
A3	2550	23	10	1550°F.	25.5
Av.	2550	23	10	1550°F.	23.0
367 A1	2500	2 hrs.-16 min.	2 hrs.	1550°F.	33.0
A2	2500	2 hrs.-16 min.	2 hrs.	1550°F.	27.5
A3	2500	2 hrs.-16 min.	2 hrs.	1550°F.	26.0
Av.	2500	2 hrs.-16 min.	2 hrs.	1550°F.	29.0
352 A1	2500	2 hrs.-21 min.	2 hrs.	1550°F.	30.5
B1	2500	2 hrs.-21 min.	2 hrs.	1550°F.	29.0
B2	2500	2 hrs.-21 min.	2 hrs.	1550°F.	28.5
B3	2500	2 hrs.-21 min.	2 hrs.	1550°F.	28.5
C1	2500	2 hrs.-21 min.	2 hrs.	1550°F.	31.0
C3	2500	2 hrs.-21 min.	2 hrs.	1550°F.	30.0
Av.	2500	2 hrs.-21 min.	2 hrs.	1550°F.	29.6
353 A1	Not heated	.....	.....	1550°F.	27.0
A3	Not heated	.....	.....	1550°F.	31.0
B1	Not heated	.....	.....	1550°F.	27.5
B2	Not heated	.....	.....	1550°F.	29.5
B3	Not heated	.....	.....	1550°F.	29.0
C1	Not heated	.....	.....	1550°F.	29.5
C3	Not heated	.....	.....	1550°F.	27.5
Av.	Not heated	.....	.....	1550°F.	28.7

Chemical analysis of the steel:

Carbon	0.47 %
Manganese	0.67 %
Chromium	0.74 %
Nickel	1.20 %
Sulphur	0.018%
Phosphorus	0.011%

Fig. 22 shows the relation between per cent reduction of area and temperature of excessive heating previous to annealing on the plain carbon steels. For the S.A.E. 1015, 1030 and 1085 steels the highest temperature plotted is just below the burning point, whereas, the S.A.E. 1050 steel was heated above the burning point, which for this steel is 2575 degrees Fahr. It will be observed that the reduction of area does not alter so long as the burning temperature of the steel is not exceeded. When the burning point is exceeded, as is the case with the S.A.E. 1050 steel, a decided drop in the reduction of area results.

Fig. 23 presents similar data for the S.A.E. 3145 steel. With this steel there is a gradual decrease in the reduction of area beginning at 2450 degrees Fahr. which is 120 degrees Fahr. below the burning point of this steel. The curve for the S.A.E. 1050 steel is



repeated in order to show graphically the difference in behavior of these two steels.

### CONCLUSIONS

Every effort has been made to measure any changes that take place in steel due to heating it to temperatures just below the burning point or within a range of 400 degrees Fahr. below the burning point. Tensile tests, hardness tests, impact tests, electrical conductivity tests, microscopic examinations, McQuaid-Ehn tests, macroscopic examinations, and fracture examinations have been made in an effort to determine to what extent if any so-called overheating or excessive heating affects the properties of the steels studied after they have been normalized or heat treated by quenching and tempering. As a result of this work the following conclusions seem justified:

1. In the lower carbon plain carbon steels no effect whatever from excessive heating could be observed after proper annealing so long as the burning temperature was not exceeded.
2. The tests of the medium carbon steels showed them to be rather brittle after being heated to 2550 degrees Fahr. and subsequently annealed and heat treated. It will be observed that this temperature is close to the temperature of burning for this steel which is slightly above 2575 degrees Fahr. The appearance of the microstructure of these samples would indicate that they are in the first stages of being burned. Although they contained no intercrystalline voids, they did contain nonmetallic inclusions in the grain boundaries.
3. In the range of plain carbon steel of about 0.90 per cent carbon the steels after proper annealing or heat treatment appeared to be unaffected by the excessive heating so long as the steel was not heated above the burning point.
4. The S.A.E. 3145 alloy steel seemed to have an overheating range which extends down to 2475 degrees Fahr. Its burning temperature is slightly above 2570 degrees Fahr. Heating this steel into the range from 2475 degrees Fahr. to 2570 degrees Fahr. and subsequently normalizing or heat treating it seemed to produce a gradual decrease in ductility with increase in temperature and to change the appearance of the fracture.

These conclusions are presented more concretely in tabular form in the following table. This table shows the maximum temperatures to which the steels listed may be heated without danger of overheating.

Impact  
Values  
(Ft. Lbs.)

28.5
23.5
30.0
27.3
23.0
20.5
25.5
23.0
33.0
27.5
26.0
29.0
30.5
29.0
28.5
28.5
31.0
30.0
29.6
27.0
31.0
27.5
29.5
29.0
29.5
27.5
28.7

of area  
on the  
steels the  
hereas,  
which  
the re-  
ture of  
d, as is  
duction

With  
begin-  
ow the  
steel is

ing or burning. These values will not hold if the furnace atmosphere is an oxidizing turbulent one, and may not hold if the period of heating is over two hours.

Type of Steel	%	Composition of Steel			Maximum temperature at which steel may be heated if overheating and burning are to be avoided. Degrees Fahr.
S. A. E. No.	Carbon	% Nickel	% Chromium	% Manganese	
1015	0.17	None	None	0.58	2650
1030	0.30	None	None	0.65	2600
1050	0.52	None	None	0.64	2525
1090	0.88	None	None	0.34	2375
3145	0.44	1.20	0.74	0.64	2450

In the writer's experience it is possible to heat just below the burning temperature, and by proper forging produce a satisfactory perfectly normal steel. However, if the temperatures given in the above table are not exceeded, one can be sure of obtaining a satisfactory product even if the steel is not forged following the heating. It is, of course, assumed that the steel will be normalized or heat treated following the high temperature exposure.

#### ACKNOWLEDGEMENTS

The writer is indebted to C. P. Teeple, who carried out many of the tests presented in this paper, and to the American Gas Association whose financial support made this study possible.

#### DISCUSSION

**Written Discussion:** By O. W. Ellis, director of metallurgical research, Ontario Research Foundation, Toronto, Canada.

Mr. Jominy is to be congratulated on having brought forward such clear evidence on the effect of atmosphere on the burning of steel. That many of the effects observed in burned steel are not concomitant with oxidation has been known for many years, but this does not in the least detract from the immense value of Mr. Jominy's contribution to our knowledge, which consists in showing conclusively that serious damage can be done to low and medium-carbon steels by heating in air to high temperatures below the solidus.

In his paper Mr. Jominy refers to the evidence presented by Stansfield and Stead in support of the belief that burning is caused by heating steel above the solidus. He then calls attention to the fact that the position of the solidus of the iron-carbon system is still in considerable doubt and shows that if a 1.1 per cent carbon steel can be considered, the solidus, according to Carpenter and Keeling, is about 230 degrees Fahr. above that according to Gutowsky. The writer has himself had occasion to question the work of Gutowsky and, in order that his ideas on this subject may be

made more clear, has prepared Fig. 1 which embodies three separate diagrams.

In the first place, there is shown in Fig. 1 the solidus of the iron-carbon diagram, according to Carpenter and Keeling. This is represented by the fine line extending from the melting point of iron to 1145 degrees Cent. (2095 degrees Fahr.) at 1.9 per cent of carbon. This line does not

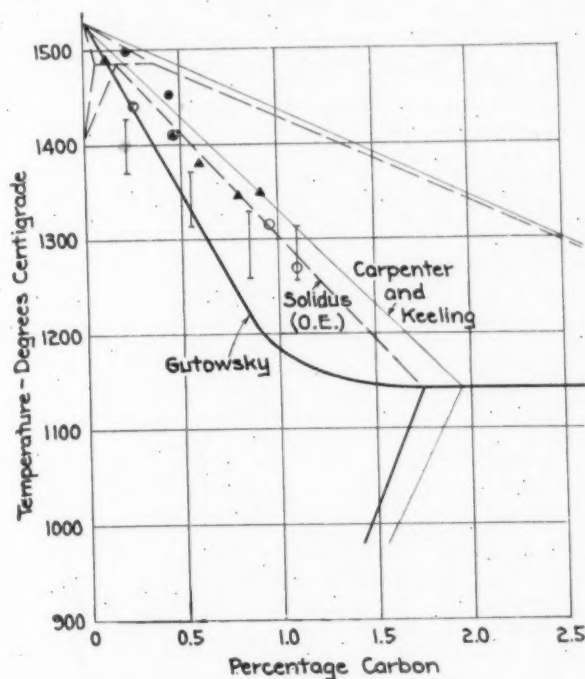


Fig. 1

coincide exactly with that shown by Carpenter and Keeling in their paper, its position having been changed somewhat to bring it into line with what has later been discovered in regard to the melting point of iron.

In the second place, there is shown the solidus of the iron-carbon diagram, according to Gutowsky. This is represented by the heavy black line which lies well to the left of the light black solidus of Carpenter and Keeling.

Intermediate between these two lines is a dotted line representing the solidus for this system, which the writer believes is more likely to be correct than that of Gutowsky. It cannot be said that the position of this line is any less certain than that of the solidus of Carpenter and Keeling, seeing that a careful examination of Carpenter and Keeling's results might lead one to draw a solidus the upper part of which, at any rate, would fall almost exactly upon the dotted line shown in Fig. 1. It will be noted that the dotted line of Fig. 1 extends from the intersection of the solidus with the peritectic line at 1486 degrees Cent. (2710 degrees Fahr.) (0.18 per cent carbon) to the intersection of the solidus with the eutectic line at the point of maximum solubility of carbon in gamma-iron (1135 degrees

Cent.—1.7 per cent carbon). This brings it into agreement with Ruer and Klesper's results in so far as the upper part is concerned, and with Gutowsky's results in so far as the solubility of carbon in gamma-iron is concerned.

The writer believes that the dotted line represents more closely than does Gutowsky's solidus the actual conditions existing in these alloys at high temperatures. He bases his belief upon the fact that when suitable samples of these alloys are forged at various temperatures, and the relationship existing between their malleability and temperature is plotted, the malleability-temperature relationship undergoes a definite alteration at temperatures represented by points very closely approximating to this line. In Fig. 1 are shown the temperatures at which a marked change in malleability occurred in five different steels tested by the writer and referred to in his Carnegie Scholarship Memoir on the "Effect of Constitution on the Malleability of Steel at High Temperatures." These temperatures are indicated by triangles, and it will be noted that in nearly all cases these triangles lie in close proximity to the dotted line. Had the marked change in malleability occurred at temperatures approaching those of the Gutowsky solidus, the writer would have been prepared to accept the Gutowsky solidus as being correct. As he pointed out in the Memoir referred to above, the method which he adopted of determining the solidus could be so improved as to be made of use in the accurate determination of the solidus of pure alloys of this or other systems.

If Mr. Jominy intends to pursue his study of the malleability of steel at high temperatures, the writer would suggest that he investigate very carefully the phenomenon of the sudden change in the malleability of steel, which occurs when steel is heated to a temperature just in excess of the solidus. The question of time of heating may be of considerable importance in this connection. The writer, however, questions whether heating, no matter how prolonged, is likely to cause the position of the solidus to be sufficiently lowered as to fall in line with that shown by Gutowsky. It seems more probable that the solidus would be raised.

On this diagram we have marked certain of Mr. Jominy's observations. Some are indicated by vertical lines, others by black dots. The former represent the ranges of temperature within which burning apparently can take place when steel is heated in an oxidizing atmosphere. It may be noted that burning did not take place at temperatures represented by the lower ends of these lines but did take place at temperatures represented by the upper ends of these lines. This diagram brings out clearly the fact that burning takes place more readily in low-carbon steels below the solidus than in high-carbon steels. The fact that high-carbon steels behave in a somewhat different way from low-carbon steels was referred to by Stansfield in his paper, and his observations in this regard are well worth repetition. He points out that:

"Professor Sexton in his 'Metallurgy of Iron and Steel' considers that oxidizing gases penetrate along the intercrystalline joints, and gives a photograph of burnt 0.3 per cent carbon steel in support of this suggestion. The passage of oxidizing gases along the joints would be more likely to



occur in low carbon steels than in those which have been mainly studied in this report. In high carbon steels above the burning-point the joints are filled with highly carburized liquid which, as long as it remained, would entirely stop the penetration of oxygen, and no appreciable burning of the steels experimented with appears to have taken place below the temperature of partial fusion, but in very low carbon steels it seems quite possible that some such action might take place at a temperature somewhat below that of incipient fusion."

It will be seen that Stansfield quite clearly admits the possibility of burning occurring at temperatures below the solidus.

The black dots indicate temperatures at which certain of these steels examined by the author showed signs of fusion. These temperatures are somewhat in excess of the solidus as indicated by the dotted line and tend to confirm the accuracy of this line rather than that of Gutowsky. However, it does seem surprising that in case of the 0.22 per cent carbon steel it should have managed to retain its shape at all at so high a temperature as 1497 degrees Cent. (2725 degrees Fahr.)

Mr. Jominy's observations in regard to the contamination of thermocouples are of considerable interest. The writer met with certain of the troubles to which Mr. Jominy has referred in his work on forging. Contamination of thermocouples when working in these high ranges is extremely difficult to prevent, and constant calibration of thermocouples is to be recommended as a precaution against error in observation.

The fact that iron and steel in the solid state will actually become hotter than the furnace atmosphere in which they are placed, when they are subjected to oxidizing conditions, is of course well known. The phenomenon was used to good purpose in days of yore and, for all I know, may even be employed today. Dr. Percy in his "Metallurgy" refers to one of the ways in which the oxidation of iron and steel was made to serve a useful purpose:

"Nailmakers, or, as they are termed, nailers, in some localities avail themselves of the combustibility of iron at high temperatures to keep their nails hot during the process of forging. There is a little blow pipe connected by a tube with a small pair of double bellows fixed above, and by this means a jet of air is thrown on the iron when it has become too cold, and immediately the temperature is much increased." Dr. Percy also notes that, "if a rod of iron be tied by a piece of cord at one end, heated to whiteness at the other, and then swung around rapidly, a brilliant circle of sparks will be produced by the combustion of the metal, resembling a firework."

H. P. EVANS: There is one question that many of us are interested in. While there is an increasing amount in tonnage every year of forgings that are being heat treated, still there are many thousands of tons of forgings that are not being heat treated, and I am wondering if Mr. Jominy has any available tabulated data regarding differences in the tensile properties without subsequent heat treatment.

**Author's Closure**

I am glad to receive the comments of Prof. O. W. Ellis whose interesting work on the malleability of steel at high temperatures has brought him in contact with the subject of burning of steel. It is rather a coincidence that the diagram which he presents in his discussion corresponds almost exactly with that presented by the present writer in Part II of the study of burning, just published. The results of this investigation seem to be in fair agreement with those reported by Prof. Ellis. I quite agree that both of our results indicate that the solidus line of Gutowsky is incorrect and that the true solidus line is fairly close to that of Carpenter and Keeling as it appears modified in most metallurgical texts and as shown in the diagram on page 390 in my paper.

I do not agree entirely with the statement of Prof. Ellis that it has been well known that iron and steel in the solid state may actually become hotter than the surrounding furnace atmosphere as brought out in my paper. Possibly Prof. Ellis' statement was prompted by the fact that it is well known that iron in the solid state will burn (chemically) in air if the temperature is sufficiently high and further that very finely divided iron will burn in oxygen at relatively cool temperatures. However that steel in the solid state should react with the relatively small amount of oxygen that may be present in the atmosphere of a forging furnace with such vigor that the temperature of the steel actually becomes higher than the furnace atmosphere has not been previously published so far as I am aware. The higher carbon steels do not exhibit this phenomenon. This is probably because they melt below the temperature at which the reaction starts, which seems to be in the range from 2475 to 2500 degrees Fahr.

This phenomenon of the steel becoming hotter than the furnace atmosphere calls attention to the need of control of atmosphere in a forging furnace. Billets and bars are often placed in the direct path of the flame in the forging furnace and even though the flame may appear to be reducing, it is possible to have shafts of air impinging on the billets or bars being heated due to the stratification of the furnace atmosphere. Under such conditions burning may occur at small sections along the bar and may pass through unnoticed. If one of these small burned areas happens to come at a point of high stress in the finished part failure is apt to result.

In reply to the last question, I should say that the results of high temperature treatments on the tensile properties would be to lower the properties as the temperature of heating was increased if there were no subsequent treatment. I do not have any data to show this but I think the effect of the high temperature treatment without subsequent heating is fairly well understood.

## STEEL INSPECTION STANDARDS FOR ROLLER BEARINGS

By A. S. JAMESON

### *Abstract*

*This paper is written in order to give the methods of inspection of S. A. E. 1015 forging quality steel used in the manufacture of roller bearings, and to show the reasons why they are adopted. It is hoped that these data may be the basis of standard specifications and metallurgical inspection methods. Macroscopic and microscopic tests and the McQuaid-Ehn test are described and illustrated. The common defects found in this forging quality steel are pointed out and their causes and prevention discussed. A short discussion is given on the production of a desirable steel for carburizing.*

### SURFACE INSPECTION

**B**ARS are visually examined for surface defects such as seams, guide scratches, overlaps, hairlines, etc. In this respect the inspection need not be rigid, for provided the seams are not deep, the upsetting does not produce defective forgings. Furthermore, the surface of the forgings is machined, thus removing all the surface defects. We find that surface defects on the bars are not as important a factor as they are when the forgings receive little or no machining. On passing the surface inspection the bars are subjected to the deep etch test.

### MACROSCOPIC EXAMINATION

The deep etch test consists of rough polishing and etching a complete cross section in a boiling solution of 50 per cent hydrochloric acid for 10 minutes. The sections are then observed for pipe, pipe being defined as undue porosity of the center of the bar; and sub-surface blow holes. On rare occasions the pipe contains large slag inclusions mainly silicates visible under the binocular microscope at low magnification. Our objection to pipe is that in the forgings

---

The author, A. S. Jameson, a member of the society, is metallurgist, West Pullman Works, International Harvester Co., Chicago. Manuscript received May 23, 1929.

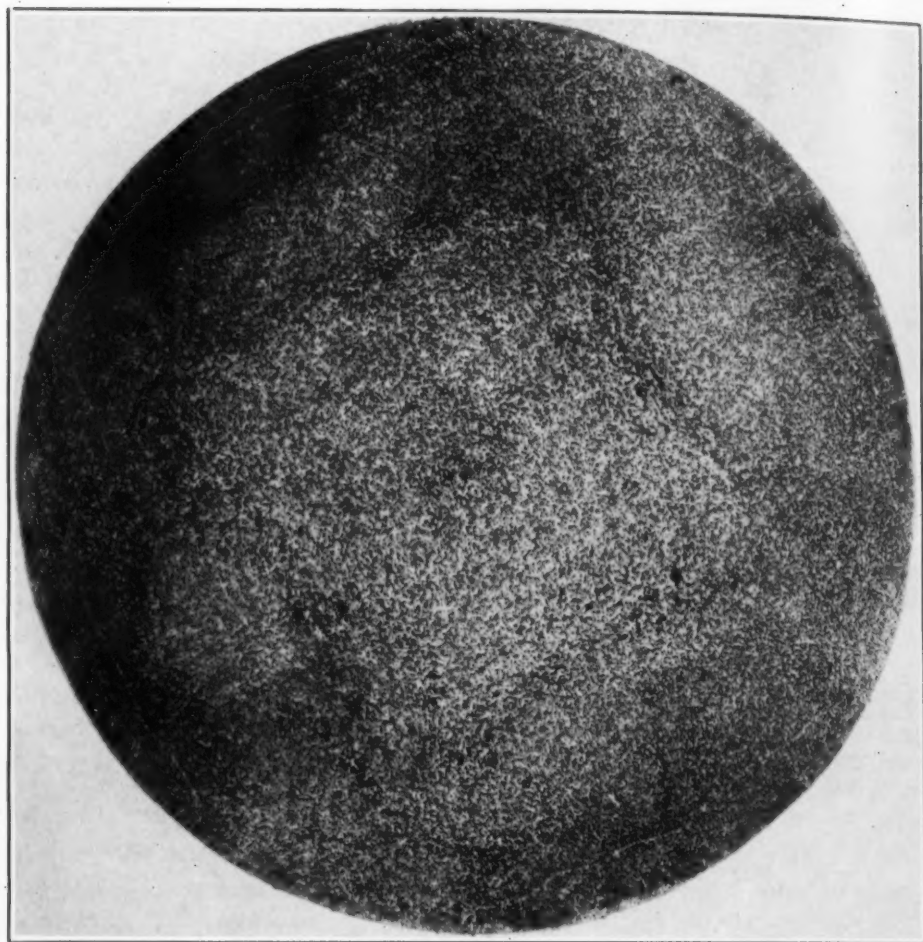


Fig. 1—Photomicrograph of Bar, Rating Good, Remarks—Slight Indication of Ingot Pattern.

the porous center becomes part of the wall of the bearing cup or cone and does not remain located in the center of the part as is the case with other types of forgings. Blow holes immediately under the surface are undesirable due to the fact that their inner surfaces become oxidized as they open up in the upsetting operation. Here they may result in defective forgings or what is infinitely worse, these oxidized blow holes may become located so far under the surface that they escape rejection in the machine shop.

The deep etch tests are made on individual furnace heats and are graded as follows:

Good or fair—pass the heat.

Poor or doubtful—acceptance dependent on other tests.

Bad—reject the heat.



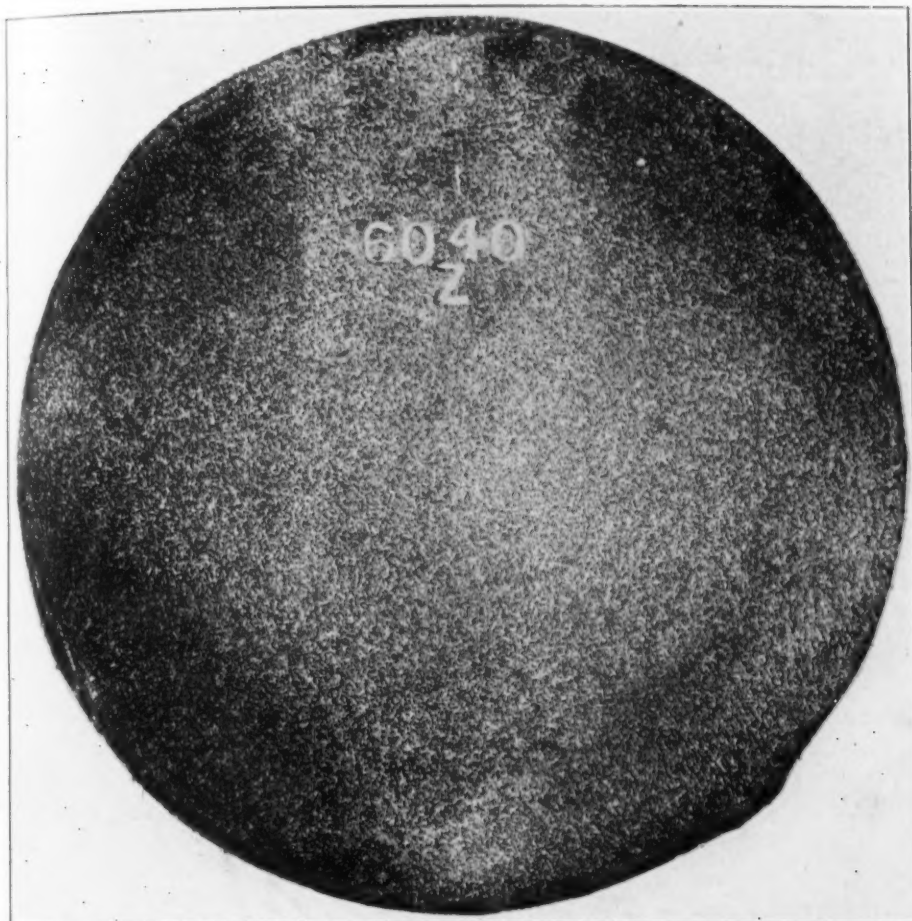


Fig. 2—Photomicrograph of Bar, Rating Good, Remarks—Slight Indication of Blowy Condition Under Surface.

Figs. 1 to 4 are part of a complete set of photomicrographs which is used by the laboratory to grade the heats. This system of grading tends to eliminate the personal element and insure the acceptance of a uniform type of steel.

*Casting Practice in Relation to Piped and Segregated Condition*

The pipe is shortened and the position of the segregate in the ingot is raised by:

1. Top casting instead of bottom casting
2. Slow casting
3. Casting with the large end up instead of down
4. Retarding the cooling of the top by means of a sinking head or otherwise
5. Liquid compression
6. Permitting deep-seated blowholes to form through adjusting the quantity of silicon and manganese or their equivalents

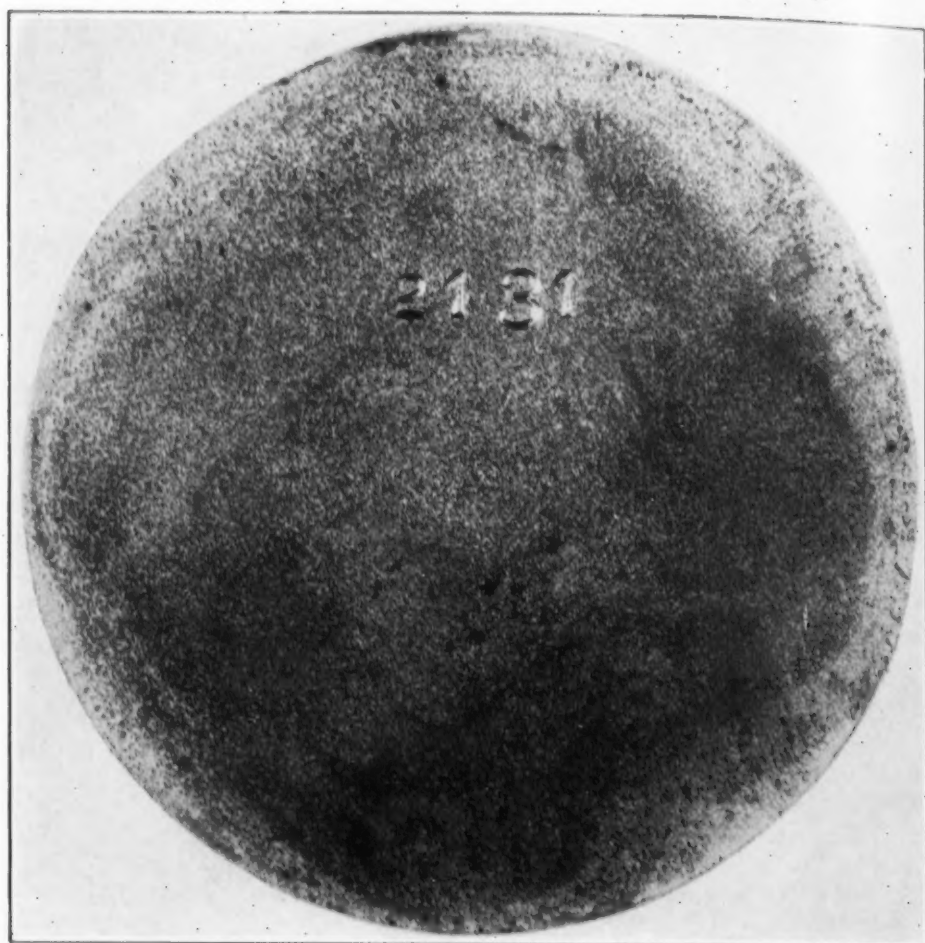


Fig. 3—Photomicrograph of Bar, Rating Bad, Remarks—Numerous Blowholes Under Surface.

The pipe is shortened and segregation increased by:

1. Casting in wide ingots
2. Using molds of low conducting power
3. Using preheated molds.

Segregation is restrained by:

1. Casting as cool as practicable
2. Casting in small molds.

#### *Segregation of Silicates and Sulphides in Sub-surface Blowholes*

The fact blowhole segregation will occur in a steel ingot with a normal average percentage of sulphur or silicon is very important to the steel manufacturer or user, because if the blowholes happen to be situated comparatively near the outer surface of the ingot they involve



Fig. 4—Photomacrograph of Bar, Rating Bad, Remarks—Piped Condition at Center.

a marked concentration of these elements where it is least desired. This segregation in what may be otherwise regarded as a good steel, will undoubtedly produce red shortness and surface cracking during the rolling or forging and a poorer surface finish on the final article.

#### MICROSCOPIC EXAMINATION

Longitudinal sections are cut from the center of the bars, polished and examined, unetched, at 100 diameters of magnification for inclusions—oxides, silicates and sulphides. This is almost a duplication of the macro-examination, for whenever the bars are piped or segregated they must invariably be dirty at the center. However, this examination is of particular value in identifying the inclusions and confirming the previous macro-examination.

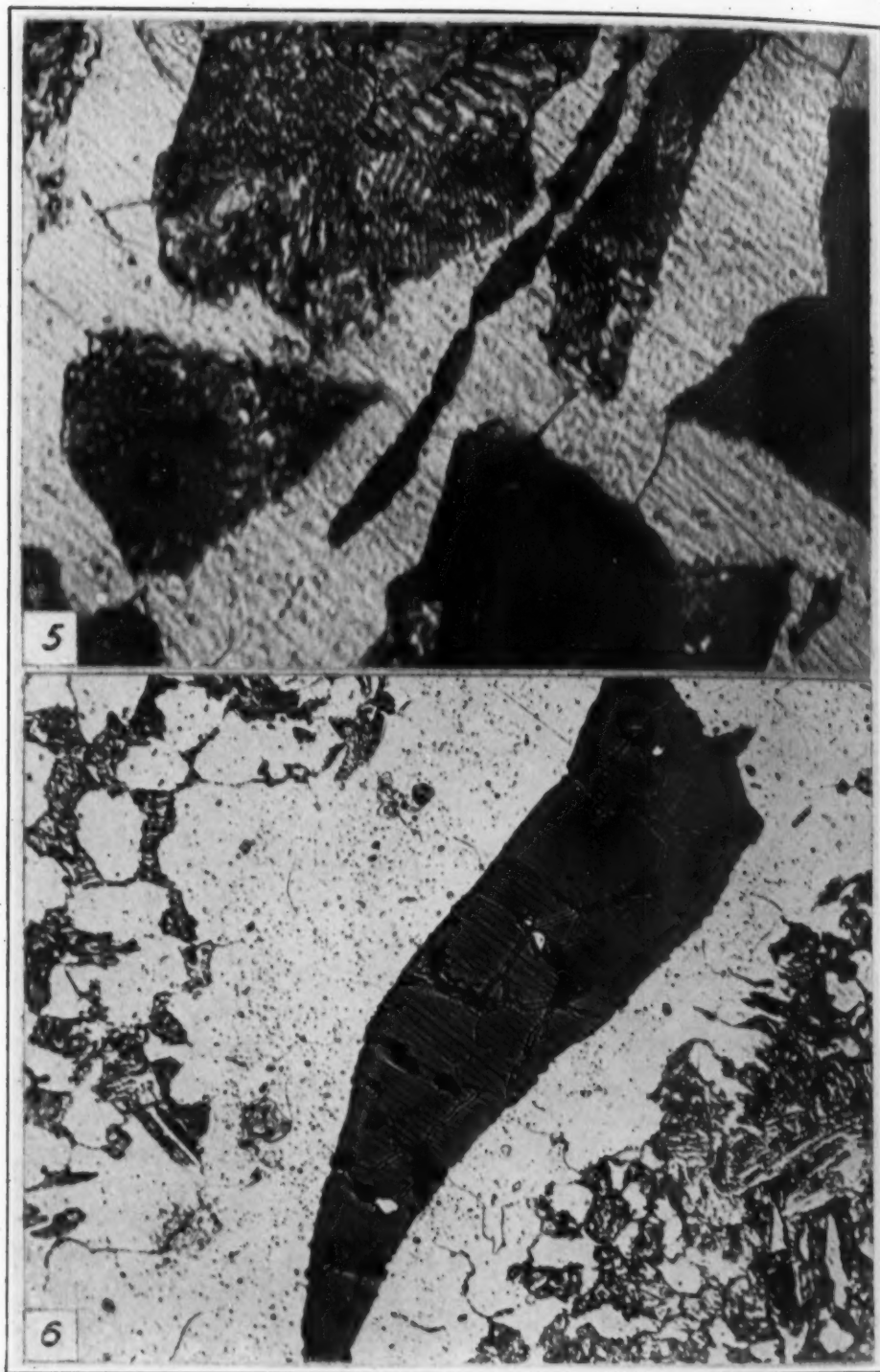


Fig. 5—Photomicrograph of Manganese Sulphide in Ferrite Band.  $\times 1000$ .  
Fig. 6—Photomicrograph of Manganese Silicate and Sulphide.  $\times 500$ . Specimens Etched with 2 Per Cent Nital.



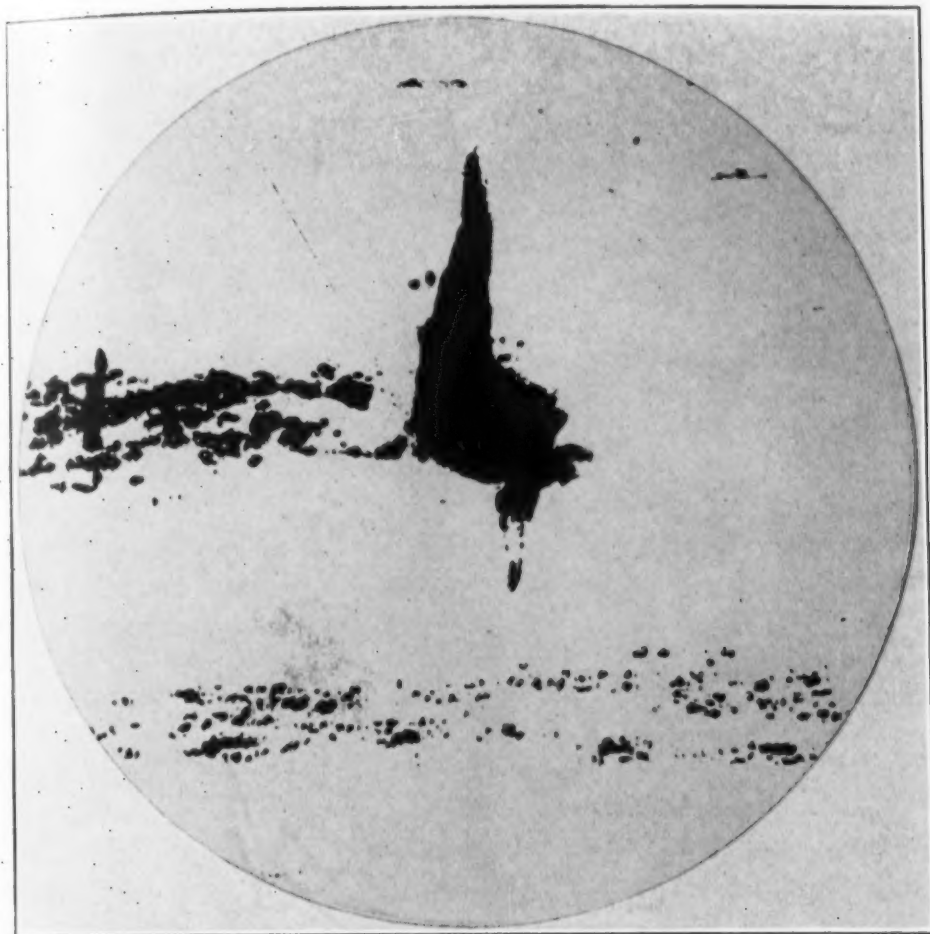


Fig. 7—Photomicrograph of Oxide Inclusions in an Unetched Specimen.  $\times 500$ .

No attempt is made to reject a heat solely on these particular samples but when the relation between this and the macro-examination is established, a clear idea of the suitability of the heat is arrived at.

Regarding the presence of nonmetallic inclusions, it seems to be well known that these are sources of weakness under fatigue stresses, precisely the conditions to which the bearing is subjected in service.

#### *Types of Inclusions*

*Sulphides* are found in all grades of steel to a greater or lesser extent being more or less an unavoidable impurity. When sulphides are situated in the interior of the bar they are less harmful than when located immediately under the surface.

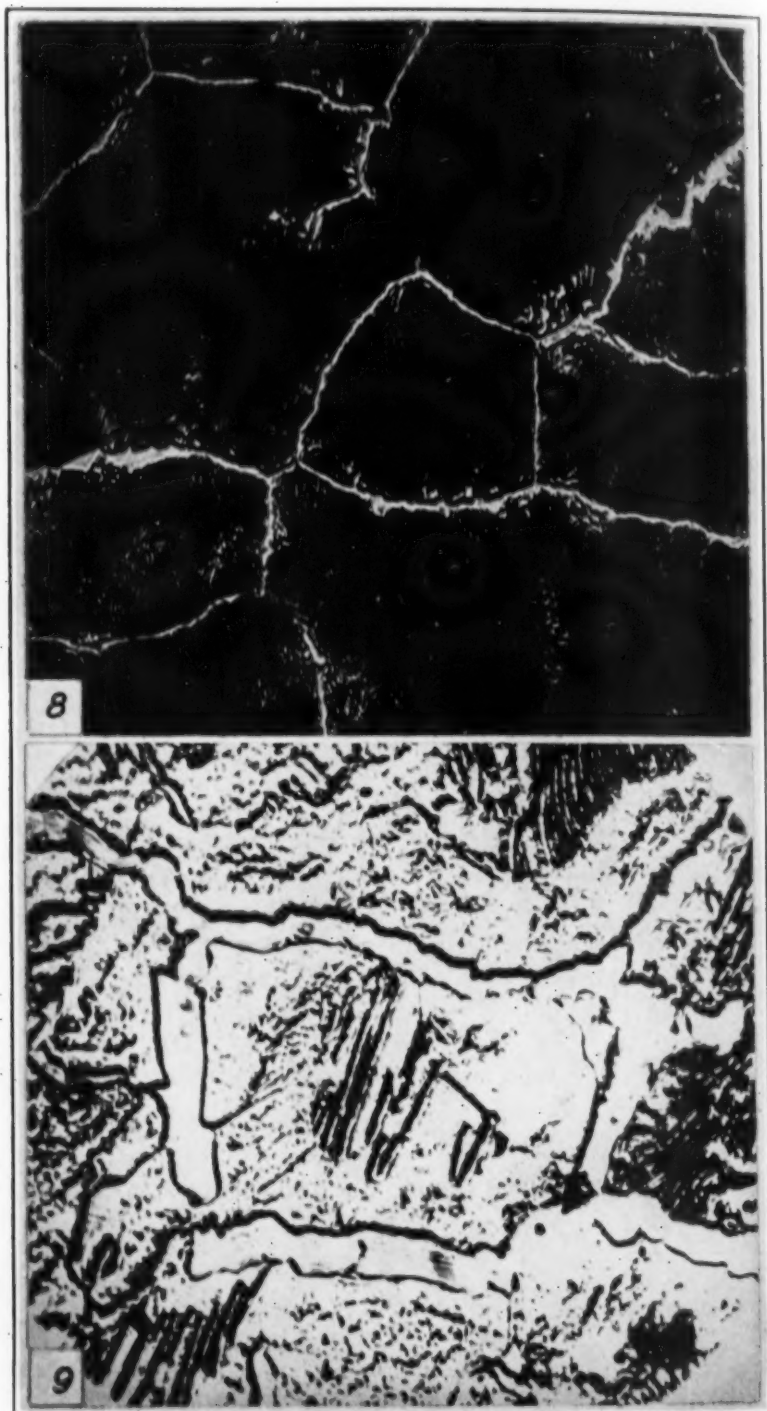


Fig. 8—Photomicrograph of Normal Case.  $\times 500$ .

Fig. 9—Photomicrograph of Abnormal Case Showing Divorced Pearlite.  $\times 1000$ .

In the presence of manganese the sulphur is present as manganese sulphide ( $MnS$ ) or  $Mn_3Fe_2S_5$ . It can be readily detected under the microscope as a dull gray or dove colored constituent and if the photograph is taken longitudinally it will be seen to be drawn out into long strings. This drawing out takes place in rolling, during which period the sulphide is plastic. On etching the sulphide is seen to be located in the ferrite bands (Fig. 5).

*Silicates* are objectionable inclusions because of their extreme brittleness, and are frequently broken and torn by the forging operation. They are found chiefly as manganese or iron silicates associated with manganese sulphide. As previously stated under the heading of macrostructure, they may be seen at low magnifications situated in the center of the bar. Silicates are decidedly darker than manganese sulphide which affords a means of distinguishing between them. From their physical characteristics which have been described and from their resultant behavior in forging we may distinguish one from the other. (Fig. 6.)

*Oxides* are found mainly as oxides of manganese, iron and alumina. (Fig. 7) To distinguish between the oxides of manganese and iron is exceedingly difficult, furthermore they are found in conjunction with silicates and sulphides.

It is difficult to draw a decided line as to the greatest number of inclusions that can be allowed because the micro sample represents such a small part of a heat of steel. However, a consideration of the size, distribution and nature of the nonmetallic inclusions will give a general idea of the conditions under which a heat was made.

### *The Effect of Inclusions*

The following effect was observed in testing a heat of pearlitic manganese steel containing 0.38 per cent carbon and 1.78 per cent manganese, used in the manufacture of automobile rear axles. The steel was normalized at 1600 degrees Fahr., quenched in water from 1550 degrees Fahr., and tempered at 1200 degrees Fahr.

Properties of Steel Tested	Yield Point in lbs. per sq. in.	Ultimate Stress in lbs. per sq. in.	Elong. in 2 inches	Reduction of area per cent
Normal properties with the above treatment	124,200 125,000	138,500 140,000	12.5 18.5	38.0 55.4

*Inclusions and Their Relation to Basic Open-Hearth Practice*

There may be said to be three types of steel, (1) rimmed, (2) semi-killed, (3) killed. Each type of steel has its own typical inclusions and segregation of inclusions. The type of steel made for roller bearings is a killed steel, the deoxidizing or killing being carried out with silicon. The silicon is added both to the furnace and to the ladle. The heat being quieted by the addition to the furnace and extra silicon being added to the ladle to complete the deoxidation as far as is desired. At the same time that silicon is added in the ladle, manganese is also added. The amounts of ferrosilicon and ferromanganese added depend upon the amount of iron oxide present in the metal at the time of tapping. The silicon and manganese combine with the iron oxide to form manganese silicate or iron manganese silicates. These silicates then are the main inclusions formed and whether they are eliminated from the metal or remain in the metal is a function of their fusibility. The more fusible they are the more rapidly will they coalesce and the larger the particles the more rapidly will they rise to the slag. The fusibility of the inclusions is governed by the amount of iron oxide making up their composition, the greater the percentage of iron oxide in them the more fusible will the inclusion be. It follows that the more silicon added as a deoxidizer the more infusible will be inclusions due to their high silica and low iron oxide content. Having produced a fusible silicate the elimination of the silicate will be considered. This is a function of the time of holding in the ladle, agitation of the steel, the temperature of the steel and the surface tension of the particle or the slag.

The open-hearth practice cannot at the present time be so closely controlled as to take into account all the factors mentioned. Two important factors however can be controlled within certain limits they are the type of inclusion formed and the temperature of the steel on pouring. It would seem that the complete or dead killing of the heat would not be advisable and that the steel should not be tapped at too low a temperature. Fortunately, this is in parallel with the production of a "normal" steel also (see the production of a desirable steel for carburizing) so that by striking a balance in the deoxidation and maintaining a pouring temperature within limits, we may obtain both a clean steel and a "normal" steel.

**McQUAID-EHN TEST**

A small-section steel sample is carburized for 8 hours at 1700



1929

to 1725 degrees Fahr. and slowly cooled. The specimen is etched with a solution of nitric acid in alcohol and the state of the carbide of the hypereutectoid zone is observed. It is required that the carbide be in the network form, which is to be fine and continuous and the pearlite show uniform distribution of the ferrite and iron carbide. A steel of this type is classified as a normal steel. (Fig. 8). By this requirement it is ensured that the carbides are completely taken into solution during the hardening heat. Steels classed as abnormal are those in which the excess carbide is massive and isolated and is, therefore, not in the network form; the pearlite is not a uniform pearlite but what is known as divorced pearlite and consists of segregates of cementite and areas of free ferrite. (Fig. 9) In such steels it is extremely hard to put the cementite into solution except by prolonged heating, the result is soft areas on water hardening, a tendency towards soft spots in brine hardening and slashes of cementite in the final product.

The test does not go as far as to take into consideration the eutectoid zone or the core. The size of the grains of the hypereutectoid zone is noted. It is desired that the grain be uniform and that the size will not vary from heat to heat.

The matter of the cause of the soft areas presents some difficulty of precise explanation here but it is in relation to the oxygen content of the ferrite which determines the state of the carbide.

The heats are classified under two headings for acceptance or rejection. The classes are normal (n) and abnormal (a). The grain size is numbered according to a grain size chart. (Grains per square inch).

- |                         |                |
|-------------------------|----------------|
| 1. Up to $1\frac{1}{2}$ | 5. 12 to 24    |
| 2. $1\frac{1}{2}$ to 3  | 6. 24 to 48    |
| 3. 3 to 6               | 7. 48 to 96    |
| 4. 6 to 12              | 8. 96 and more |

#### *The Production of a Desirable Steel for Carburizing*

It is usual to term a desirable carburizing steel a "normal" steel. A "normal" steel is one in which after carburizing and slowly cooling, the excess cementite is precipitated at the grain boundaries, and the cementite is then in continuous and fine network form around grains of pearlite.

In an "abnormal" steel the cementite is massive and isolated and surrounded by areas of ferrite. The main characteristic of "abnormal" structure is the lack of perfection of the pearlite, as seen by

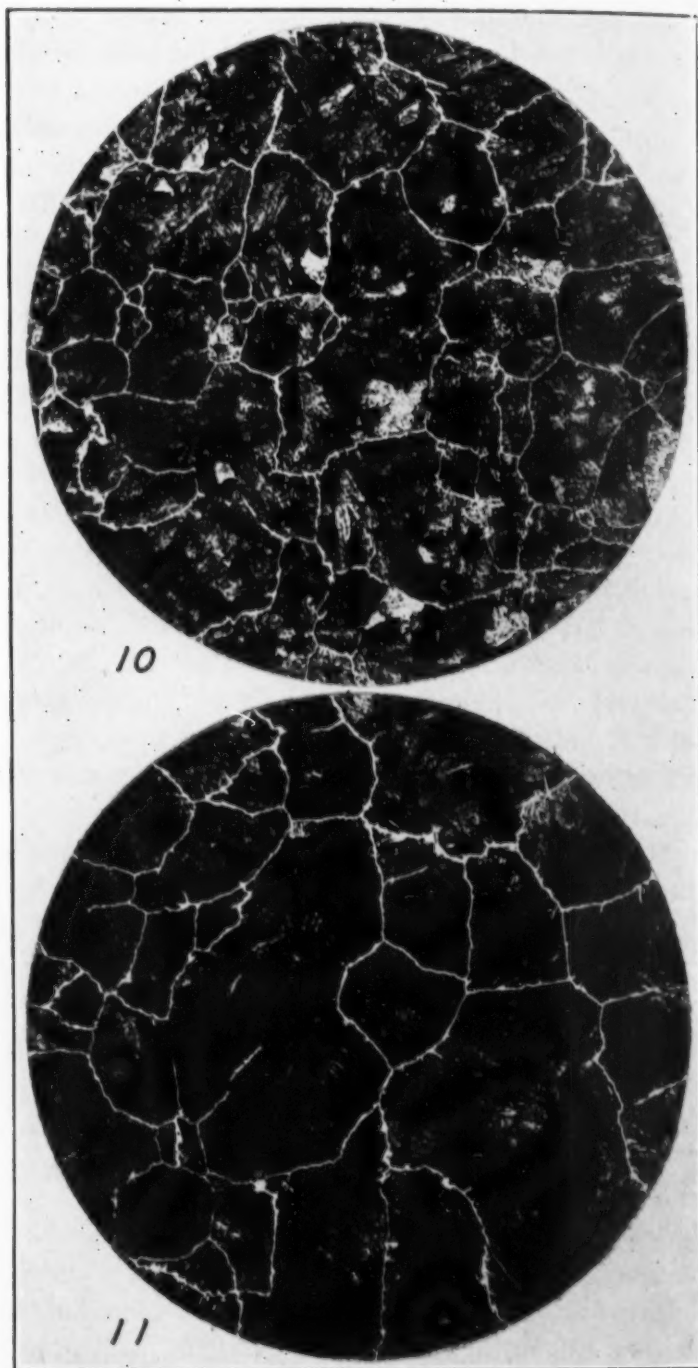


Fig. 10—Photomicrograph of Normal Case of Aluminum-treated Steel. Grain Size No. 3.

Fig. 11—Photomicrograph of Normal Case of Silicon-treated Steel. Grain Size No. 1.

the coalescence of the cementite and its separation from the ferrite. There is also a question of grain size and a balance must be attained between an extremely coarse grain and an extremely fine grain.

The advantages of a "normal" steel for carburizing are (a) the rate of carburization is greater, (b) the cementite is more readily taken into solution in the hardening heat, (c) there are no soft spots in water quenching and no tendency towards soft areas in brine quenching. The result is a greater uniformity of hardness which accrues from (b) and (c). As to grain size a coarse-grained normal steel shows an increased rate of carburization, often the increase will amount to 15 per cent—no negligible consideration.

Where the question of the cost of carburizing is of primary importance a coarse-grained steel would be advantageous. However, the coarse grain is accompanied by a certain degree of brittleness, and as carburized steels are desired to be tough, a finer grain size is required in order to get this toughness. It will be better then to produce a steel of moderate grain size.

The production of a "normal" or "abnormal" steel seems to be bound up in the amount of residual oxygen in the metal. In a "wild" heat which would necessarily contain more dissolved oxygen, the steel is abnormal in certain parts of the ingot, in others quite normal. A completely "dead killed" heat is also generally abnormal. The most convenient control may be exercised in the final deoxidation of the heat though the heat should be first shaped up in the furnace. As there must be a residual oxygen content in the metal, the use of deoxidizers in the mold is not a good practice either for the production of a "normal" steel or a clean steel. Deoxidizing is carried out in the furnace and in the ladle. As to the deoxidizers whether silicon, aluminum, vanadium, or titanium are used is a matter of choice to some extent for so long as the heat remains slightly open and not completely dead killed the steel will be "normal." Figs. 10 and 11.

The use of silicon in the ladle produces a normal steel with a coarse grain size. Aluminum is generally thought to produce an "abnormal" steel. The explanation seems to be that as aluminum is so powerful a deoxidizer that the chance of over killing the heat is more likely. If aluminum is used in small quantities in the ladle it does not produce an abnormal steel but has the advantage of reducing the grain size and increases the ingot yield whilst it is easier to roll than a silicon-killed heat.

## Educational Section

These Articles Have Been Selected Primarily For Their Educational  
And Informational Character As Distinguished From  
Reports Of Investigations And Research

### THE CONSTITUTION OF STEEL AND CAST IRON SECTION II—PART XIII

By F. T. SISCO

#### *Abstract*

*This installment, the thirteenth of the present series, takes up thermal treatment as applied to cold-worked steels. After a brief review of the annealing of cold drawn and cold-rolled products the operation of patenting is described. This process which is unique in heat treatment involves heating the material well above the critical range followed by slightly accelerated cooling in air or lead with the object of forming large grains of finely granular pearlite or sorbite. Following this, as representative of hardening and tempering applied to cold drawn wire, the operation of oil tempering is described. The effect of galvanizing upon the properties of cold drawn steel is the last subject discussed.*

A VERY large percentage of all cold-worked steels is drawn directly from the hot-rolled bar or rod, with the exception, of course, of such process-annealing as may be necessary to restore the ductility so that cold drawing can be carried further. There is however, a large class of cold drawn products known as high strength wire which, although of relatively small tonnage when compared with the total, is very important. Among these products may be mentioned spring wire, rope wire, music wire and the like. In the manufacture of this special or so-called high strength wire some form of

This is the twenty-third installment of this series of articles by F. T. Sisco. The several installments which have already appeared in TRANSACTIONS are as follows: June, July, August, September, November, 1926; January, February, April, June, August, October, 1927; February, April, June, November, 1928; January, March, May, June, July, September, October, 1929.

The author, F. T. Sisco, a director of the society, is chief of metallurgical laboratory, Air Corps, War Department, Wright Field, Dayton, Ohio.



heat treatment either before, during or after cold working may be an important factor. This heat treatment may include normalizing, patenting (sorbitizing), hardening, or tempering, either alone or in combination as may be necessary to secure the properties desired.

The heat treatment operations that are used in the wire industry and will, consequently, receive our special attention, include annealing below the critical range, patenting, and the heat treatment of spring wire which is known in the wire mills as oil tempering. In general the processes of normalizing, annealing above the critical range, hardening and tempering as applied to cold-worked steels, if used at all, are but little different in procedure than these same operations applied to hot-rolled, forged or machined sections. For example, normalizing is occasionally necessary for hot-rolled rods which are to be drawn into high strength wire, in order to remove rolling strains and the slight heterogeneity of structure that sometimes is encountered in hot-rolled steel. Galvanizing, although not strictly a process of thermal treatment is, when used for high strength wire, actually a tempering operation and has a noticeable effect upon the physical properties. For that reason it will be described briefly.

#### ANNEALING COLD-WORKED STEELS

The operation of annealing as applied to cold-worked steels has been discussed in a previous installment<sup>74</sup> where particular attention was paid to the structural changes taking place, including annealing below the critical range and the important factor of recrystallization. It will be unnecessary to do more than summarize this briefly. Annealing as applied to cold-worked steel has two objects: the first is to relieve internal strains and the second to effect complete recrystallization. For most wire one operation accomplishes both. The annealing operation is sometimes carried out below the lower critical point,  $A_{c1}$ , sometimes just at the point, and occasionally above it, depending upon the composition of the material and the physical properties desired.

One of the noticeable effects of cold working is that the grains become more-or-less severely distorted in the direction of work. Pure iron (ferrite) has a definite temperature above which recrystallization takes place. This temperature is, as we have seen, about 965 degrees Fahr. (520 degrees Cent.). Consequently if a cold drawn

<sup>74</sup>TRANSACTIONS, American Society for Steel Treating, June, 1928, page 1044.

wire of pure iron or low carbon steel is heated to 965 degrees Fahr. or above, recrystallization will take place; the elongated, distorted grains will be replaced by equiaxed grains. In the case of steels containing an appreciable amount of pearlite, heating to this temperature serves to recrystallize the ferrite grains but the elongated particles of pearlite remain unaffected.

Another well known effect of annealing after cold work is the spheroidization of the cementite which occurs at 1200 to 1290 degrees Fahr. (650 to 700 degrees Cent.) or just below the lower critical point,  $Ac_1$ . If a hot-rolled and annealed medium or high carbon steel is heated to this temperature spheroidization of the cementite takes place slowly if at all; but if this steel is cold-worked and then heated to 1250 to 1290 degrees Fahr., the spheroidization proceeds with great rapidity. The change in structure that takes place due to the divorce of the pearlite was illustrated in a former installment<sup>75</sup> where photographs were shown of a hard drawn streamline wire reheated to 1100 and 1200 degrees Fahr. These micrographs show the practically complete spheroidization which occurred in this wire after heating for one hour.

Process annealing to soften cold-worked low carbon steel sufficiently so that it can be drawn to smaller sizes is usually accomplished by heating to a temperature varying from 900 to 1200 degrees Fahr. or below the critical range. Annealing the finished wire to remove strains, reduce the strength and increase the ductility is usually carried on at the same temperature. A few examples given in the following table will illustrate this.

**Pot Annealing Practice for Common Wire**

Size Wire	Designation	Temperature Degrees Fahr.
About 0.090 in.	Rapid Annealing	1150 to 1200
Large Sizes	Rapid Annealing	1050 to 1100
0.035 in.	Common Annealing	900 to 950
0.070 in.	Common Annealing	1100 to 1150
Bale Ties	Blue Annealing	750 to 800
0.016 to 0.062	Blue Annealing	1100 to 1150
Fine Wire (very small sizes)	Black Annealing	900 to 1000
0.040 to 0.070	Bright Annealing	750 to 800
	Bright Annealing	1150 to 1200

<sup>75</sup>TRANSACTIONS, American Society for Steel Treating, June, 1928, page 1047.

The descriptive words, blue, black and bright, refer to the appearance of the wire after it comes out of the annealing pots. Needless to say in annealing bright wire special precautions are taken so that no scale will form on the wire during the heating, soaking or cooling.

Medium and high carbon steel wire if annealed at all, is process annealed by heating to the spheroidization temperature (1100 to 1200 degrees Fahr.). The complete formation of spheroidized cementite, which occurs rapidly after the wire has been cold drawn results in maximum softness and the most ductile condition for further cold work.

In general it is more difficult to anneal hard-rolled steel sheet and strip than wire, but these difficulties lie chiefly in the necessity for preventing the formation of scale and keeping the sheets from sticking to each other. The temperatures used vary from 1200 to 1400 degrees Fahr.

#### THE PATENTING OF WIRE

Among the heat treatment operations carried on to improve the properties of cold drawn wire is the unique process of patenting. Briefly, this consists of heating the hot-rolled rods<sup>76</sup> of medium or high carbon steel to a temperature several hundred degrees above the lower critical point, followed by slightly accelerated cooling. The object of patenting or sorbitizing, as it is sometimes called, is to produce a characteristic structure in the rod consisting of very large grains of fine granular pearlite or sorbite. This is produced by the combination of overheating and accelerated cooling.

In a previous installment<sup>77</sup> we saw that the operation of patenting consisted of passing the wire continuously through a long furnace. The cold wire enters at one end and is gradually heated until when it emerges from the other end it is at the desired temperature. This is one of the few operations in steel making that has been developed to a high state of perfection independent of metallurgical control. The difficulties involved in patenting become readily apparent when it is realized that the grain size of the properly patented wire must be large but not too large; the size depending upon how much working the wire receives and what properties

<sup>76</sup>In wire manufacture steel billets are hot-rolled to small round bars varying between 0.200 and 1.00 inch in diameter. These are known as rods.

<sup>77</sup>TRANSACTIONS, American Society for Steel Treating, June, 1928, page 1053.

are required; and in addition the structure of each grain must be a fine granular pearlite or sorbite containing little or no laminated pearlite, free ferrite or troostite. When it is remembered that patenting is a continuous process the difficulties involved in producing this structure become plain. It is obvious that the rate at which the wire travels must be carefully regulated for rods of different sizes and the temperature must be changed if necessary for steels of different analysis. The length of time that the moving wire is maintained at the desired temperature must be closely controlled or grain growth will proceed too far. When these variables are all considered, the perfection of results obtained in modern wire mills is astounding.

It has been repeatedly emphasized in this series of articles, as it is emphasized in all text books for that matter, that in heat treatment great care should be exercised not to overheat the material because grain growth of the austenite would take place. When this coarse-grained austenite is cooled it would transform into a coarse-grained ferrite, pearlite, sorbite, or martensite as the case may be. This coarse grain resulting from overheating would detract seriously from the quality of the steel, making it brittle, hence at every stage of steel manufacture, and especially in heat treatment care is taken to prevent overheating. In wire drawing, instead of being careful not to overheat the steel, this is done deliberately.

The process of patenting consists, as we have seen, of heating the wire to a temperature well over the critical range, between 1500 and 1950 degrees Fahr. depending upon the composition, holding just long enough for the required grain growth to take place followed by a sufficiently rapid cooling to form sorbite. The cooling is usually in air; for some medium and high carbon steels, however, the wire passes from the hot furnace directly into molten lead maintained at 900 to 1150 degrees Fahr. This is known as lead patenting.

Now why does the wire manufacturer deliberately overheat steel which is to be drawn into high strength wire? Because this overheated steel even though brittle, draws better, and produces a more ductile and at the same time a stronger wire. The brittleness due to overheating is practically eliminated in the first draft. It is, of course, necessary to use great care in drawing this overheated wire until the brittleness is gone. The reason why a coarse grain is desirable is this: It is generally accepted that the longer the fiber in a cold drawn wire the better it will withstand the torsion test. A patented rod with



st be a  
minated  
at pat-  
oducing  
hich the  
nt sizes  
of dif-  
s. main-  
or grain  
sidered,  
ounding.  
icles, as  
at treat-  
erial be-  
hen this  
coarse-  
may be.  
seriously  
ry stage  
is taken  
careful

heating  
een 1500  
holding  
place fol-  
cooling is  
however,  
ad main-  
patenting.  
heat steel  
his over-  
es a more  
ss due to  
of course,  
until the  
esirable is  
old drawn  
rod with

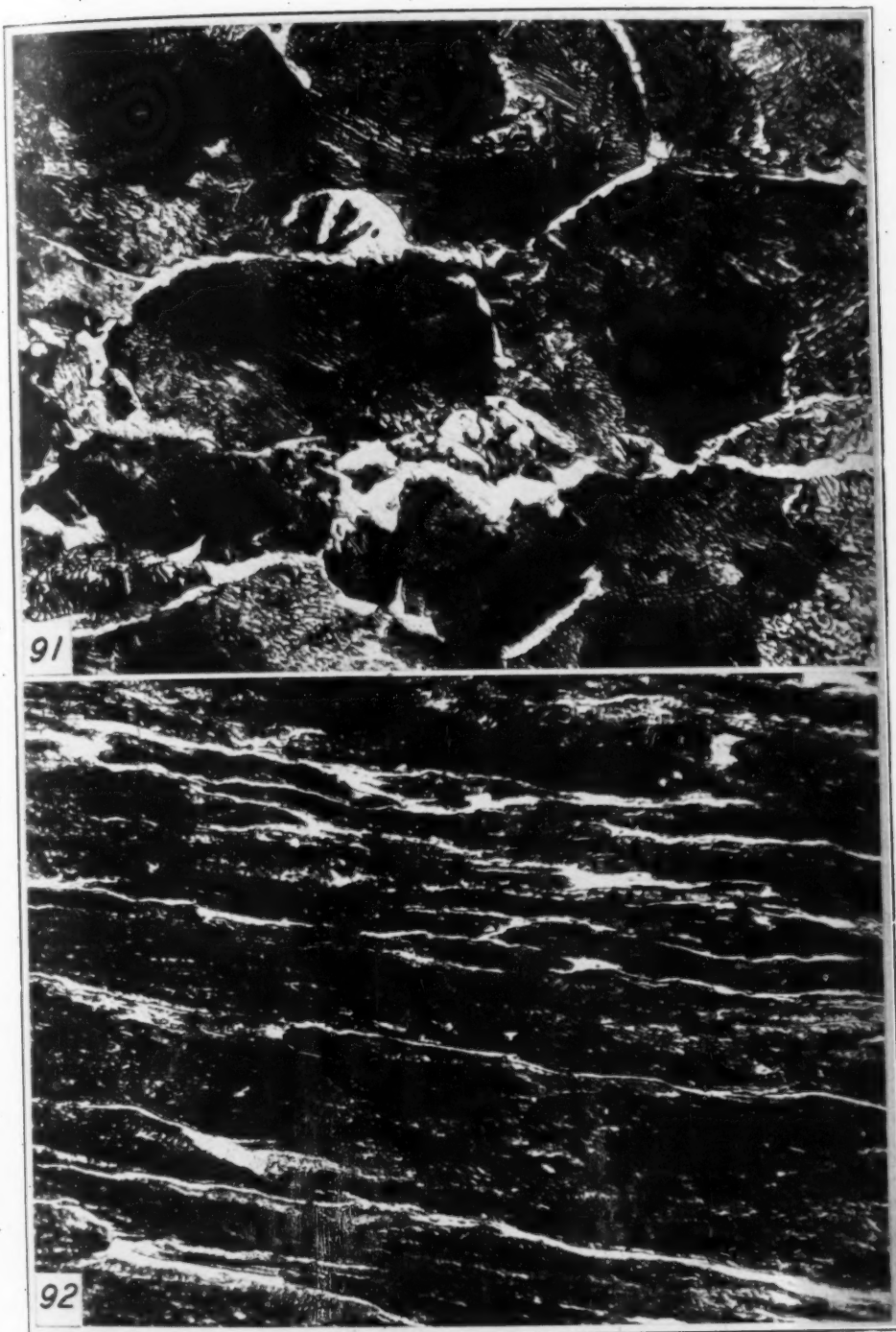


Fig. 91—Large Equiaxed Pearlite and Sorbite Grains in the Threaded Section of a Streamline Wire.  $\times 500$ . Specimen Etched with Alcoholic Nitric Acid.  
Fig. 92—Same Wire as Fig. 91. Large Elongated Grains in the Streamline Section.  $\times 500$ . Specimen Etched with Alcoholic Nitric Acid.

large grains will draw into a wire having long fibers. This is clearly shown<sup>78</sup> in Figs. 91 and 92. The former is the structure of the threaded section of a streamline wire, while the latter is the structure of the swaged portion of the same wire. It is also generally accepted that a wire with a relatively short fiber will withstand the bend test better than one with a very long fiber. This short fiber is produced by cold working a rod with a somewhat smaller grain. An example of a streamline wire in this condition is shown in Figs. 93 and 94. It is not always easy to decide whether a short fiber or a long fiber wire is best; many wire manufacturers compromise and aim at an average between the two.

The next question that concerns us relative to the patenting process is why a sorbitic structure is so desirable. Adam<sup>79</sup> gives a clear example of this. An 0.85 per cent carbon steel rod was annealed at 1650 degrees Fahr. and cooled slowly in the furnace. Another specimen was patented at 1925 degrees Fahr. and cooled in lead at 1100 degrees Fahr. The annealed wire broke when drawn two drafts to 10-gage wire; the patented specimen drew eight drafts to 17½-gage without failure. Annealed wire in the pearlitic condition is much inferior as regards its drawing properties than wire in which the structure is wholly sorbite. The sorbitic steel is much harder but is much more ductile as measured by its capacity to withstand cold deformation.

The reason for the great difference in drawing properties of a pearlitic steel as against a sorbitic steel is as follows: In the pearlitic condition the laminations of ferrite and cementite are extremely haphazard, varying widely in different grains. Thus in many grains the relatively large, hard, brittle plates of cementite are at right angles to the direction of flow and thus oppose it. In the case of sorbitic structure the particle size is very small, the cementite and ferrite are practically emulsified, hence slip occurs readily in all directions.

Lead patenting produces a more uniform structure than in the ordinary air-cooled product. This is due to the fact that it is exceedingly difficult to cool the wire in air without the formation of a little pearlite (if cooled too slow) or a little troostite (if cooled too fast). As representative of the physical properties of patented and

<sup>78</sup>Fig. 91 has much more laminated pearlite present than is usually desired in patented wire.

<sup>79</sup>Loc. Cit., page 83.

is clearly  
e threaded  
ure of the  
epted that  
test bet-  
duced by  
sample of  
94. It is  
fiber wire  
n average

patenting  
79 gives a  
l was an-  
ace. An-  
cooled in  
en drawn  
ght drafts  
tic condi-  
n wire in  
is much  
y to with-

rties of a  
he pearli-  
extremely  
ny grains  
right an-  
se of sor-  
e and fer-  
all direc-

an in the  
it is ex-  
ation of a  
cooled too  
ented and

l in patented

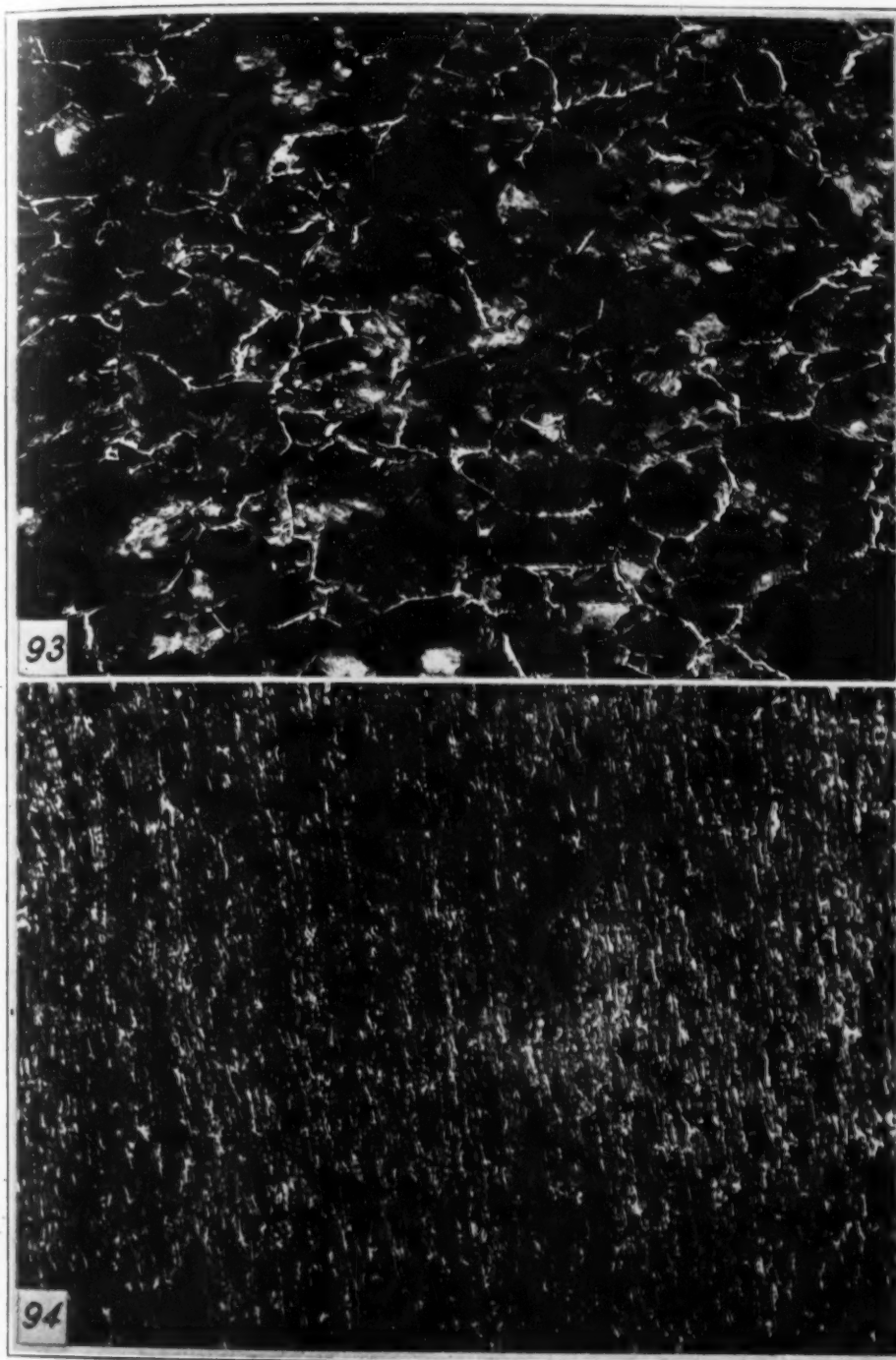


Fig. 93—Smaller Equiaxed Sorbite and Granular Pearlite Grains in Threaded Section of a Streamline Wire.  $\times 500$ . Specimen Etched with Alcoholic Nitric Acid.

Fig. 94—Same Wire as Fig. 93. Small Elongated Grains in the Streamline Section.  $\times 500$ . Specimen Etched with Alcoholic Nitric Acid.

drawn wire the following table from Adam is typical. The material was a 0.44 per cent carbon, 0.82 per cent manganese steel patented at about 1900 degrees Fahr. Much better physical properties can be obtained with higher carbon steels.

**Physical Properties of Patented and Cold Drawn Wire (Adam)**

Per Cent Reduction Per Pass	Diameter Inches	Ultimate Strength Pounds Per Sq. In.	Air-Cooled		Per Cent Reduction Per Pass	Diameter Inches	Lead-Cooled Ultimate Strength Pounds Per Sq. In.	Torsions	
			Torsions in 100 Diam.	Bends				in 100 Diam.	Bends
none	0.160	110,000	..	..	none	0.160	116,500	..	..
24	0.139	130,000	35	6	23	0.140	130,000	25	6
24	0.121	143,400	31	7	25	0.121	145,600	28	8
30	0.101	156,800	37	7	20	0.108	150,000	32	10
26	0.087	172,500	38	11	22	0.095	163,500	33	12
30	0.073	183,700	39	16	31	0.079	179,200	38	15
30	0.061	197,100	38	19	40	0.061	199,400	40	24
35	0.049	224,000	38	18	40	0.047	235,000	38	28
33	0.040	244,200	35	18	35	0.038	275,500	36	34
46	0.030	302,400	31	17	38	0.030	315,800	34	34

#### THE HEAT TREATMENT OF WIRE

The thermal treatment of wire, including hardening and tempering, is carried on for much the same reason that hot-rolled or machined products are heat treated—to improve the physical properties. One of the most common treatments applied to high strength wire is the so-called oil tempering. As this process is typical of the heat treatment of wire products it will be described briefly. The wire is passed continuously through a long narrow furnace where it is heated to the proper hardening temperature. It then moves into a long trough containing the quenching oil. This oil flows in at one end, out the other and then through a cooling system. From the oil quenching bath the wire passes into a lead tempering bath. The so-called oil tempering is a combination of hardening by quenching in oil followed by tempering in molten lead between 800 and 1100 degrees Fahr.

The physical properties of oil-tempered wire range from a minimum of about 170,000 pounds per square inch tensile strength and an elongation of 10 per cent in 10 inches for wire one-half inch in diameter to over 250,000 pounds per square inch tensile strength and 1.00 per cent elongation for wire 0.020 to 0.050 inch in diameter. Oil-tempered wire is used extensively for springs. The average tensile strength of oil-tempered wire for springs is 180,000 to 200,000 pounds per square inch with 3.00 to 6.00 per cent elongation.



The ma-  
ese steel  
cal prop-

Adam)

Torsions in 100 diam.	Bends
25	6
28	8
32	10
33	12
38	15
40	24
38	28
36	34
34	34

and tem-  
-rolled or  
ical prop-  
a strength  
typical of  
effly. The  
e where it  
ves into a  
in at one  
om the oil  
The so-  
nching in  
1100 de-

m a mini-  
length and  
lf inch in  
length and  
diameter.  
verage ten-  
o 200,000  
on.

Most of the steel wire which is galvanized is low carbon and a large percentage of it is desired soft, with relatively low tensile strength and high ductility. There are a few products, however, which are made of high strength galvanized wire. Among these are rope and cable for use out of doors. For this reason we will look briefly at the effect of galvanizing on the strength and ductility of hard drawn high strength wire. Galvanizing takes place at a temperature of 900 to 950 degrees Fahr. As a rule the wire passes rapidly through the zinc kettle, sufficiently fast so that in many cases it does not get heated completely through its cross section. It does, however, get hot enough so that the strength and ductility of the wire are oftentimes seriously affected. Galvanizing as ordinarily practiced has a deleterious effect in that it makes the wire brittle. The cause for this is not known, but the effect of this temperature (900 to 950 degrees Fahr.) on the properties has been thoroughly studied.

Passing a high strength wire through a galvanizing kettle reduces the ultimate strength and increases the elongation to a marked degree. But although the elongation is increased the brittleness as measured by the torsion and bend test is also increased. The reduction in torsion of a 0.60 per cent carbon wire due to galvanizing is as much as 30 per cent; before galvanizing wire which will withstand 44 torsions, afterwards will withstand only 30. For higher carbon steels the increase in brittleness is even greater. Adam found that galvanizing an 0.80 per cent carbon wire reduced the torsion value 63 per cent. In general then, we may say that galvanizing has a harmful effect upon the physical properties of high strength wire, making it more brittle. This characteristic should be taken into consideration when specifying that high strength wire should be galvanized.

## Recommended Practice Committee Releases\*

### TENTATIVE RECOMMENDED PRACTICE FOR A STANDARD MACRO ETCH TEST FOR IRON AND STEEL†

**GENERAL**—Macro etching of iron or steel consists of subjecting the metal to the action of an acid to bring out its structure for visual inspection. This recommended practice on macro etching covers the preparation of the sample, composition and the temperature of the etching solution, the time of etching, and the interpretation of the results.

**Preparation of Sample**—When it is desired to reveal the surface defects, such as seams, laps, and grinding checks, preparation of the sample is frequently unnecessary. Scale on the surface of the material will be removed by the acid, so that the surface defects will become visible, but in order to reveal the internal structure or defects in a bar or forging, a section must be prepared by cutting through the metal at a point where inspection is desired.

The surface finish of the specimen depends on the etching solution used. A machine surface is recommended as the standard surface finish in this practice. A ground and polished surface is not necessary and in many instances a fairly smooth sawed cut is satisfactory, but for some of the etching solutions, such as Humfrey's, LeChatelier's, ammonium persulphate, etc., a well polished surface is required.

**Etching Solution**—There are many types of etching solutions which can be used, but in this practice a solution of one part commercial hydrochloric acid and one part water is recommended. This solution has the advantage that it can be heated without change in concentration. The fumes of this mixture are very corrosive so that the work is best done under a hood.

It is not necessary to throw away the acid after each test, but for the best results a fresh solution should be used for each test.

For containers pyrex glass, porcelain dishes, or corrosion resisting metals can be used. For heating the solution the method which gives the best temperature control should be preferred.

**Temperature of Acid**—A temperature of 160 degrees Fahr. is recom-

†This is a—Tentative Recommended Practice—and will remain tentative for at least one year, until adopted by the Board of Directors and the Recommended Practice Committee of the A. S. S. T. This practice is not intended for a specification and should not be interpreted as such.

The membership of the Sub-Committee on Macro Etching was as follows: H. G. Keshian, Chairman, G. J. Comstock, F. R. Palmer, G. V. Luerssen, L. A. Lanning, N. B. Hoffman, J. P. Gill, H. J. Fischbeck, J. J. Curran, M. G. Yatsevitch, and W. R. Shimer.

Criticisms of this tentative practice are solicited and should be directed to J. E. Donnellan, Secretary of the Committee, 7016 Euclid Avenue, Cleveland, Ohio.

\*The releases from the Recommended Practice Committee as printed herewith will be included in the next edition of the A. S. S. T. HANDBOOK, which will be published about June 1, 1930.

Table I  
Recommended Etching Time for Various Steels

S.A.E. Carbon Steels	Minutes	S.A.E. Nickel-Chromium Steels	Minutes	S.A.E. Molybdenum Steels	Minutes
1010	15	3115	45	4130	45
1015	15	3120	45	4140	45
1020	15	3125	45	4150	45
				4615	45
1025	30	3130	45	S.A.E. Chromium Steels	
1030	30	3135	45	5120	30
1035	30	3140	45	5140	30
1040	30	3215	45	5150	30
		3220	45	52100	45
1045	45	3230	45	S.A.E. Chromium-Vanadium Steels	
1050	45	3240	45	6120	45
1095	45	3245	45	6125	45
		3250	45	6130	45
1315	30			6135	45
1350	30	3312	60	6140	45
1360	30	3325	60	6145	45
		3335	60	6150	45
		3340	60	6195	45
S.A.E. Nickel Steels		3415	45	S.A.E. Silico-Manganese Steels	
2315	30	3435	45	9250	30
2320	30	3450	45	9260	30
2330	30	S.A.E. Tungsten Steels		Stainless Steels	
2335	30			Stainless Irons	30
2340	30	71360	45	Stainless Steels and	
2345	30	71660	45	18-8 Stainless	
2350	30	7260	45	Steels	45
2512	30				
Tool Steels					
Carbon and Carbon-Vanadium				45	
Manganese Oil Hardening				30	
Fast Finishing				45	
Tungsten Hot Die Steel				45	
Chromium Hot-working Die				45	
High-Carbon-High-Chromium				45	
High Speed				45	

mended. This temperature gives a vigorous reaction and does not evaporate the solution too rapidly. The temperature should be determined with a good thermometer.

It is recommended to first clean the specimens and then heat them in hot water to the same temperature as the acid. The specimens are then transferred to the acid bath, which has first been heated to the proper temperature. By so doing it is much easier to control the time element than if the specimens are put into cold acid and brought up to temperature with the solution, and also it is easier to duplicate conditions time after time.

**Time of Etching**—The time of etching will depend upon the type of steel to be etched. It is extremely important that the time be accurately determined. If the steel is not etched long enough, the sample will not give all of the information desired; if it is etched too long, some of the more delicate details will be masked by the general destruction of the surface.

When the specimens are in the annealed condition, the etching periods given in Table I have been found to give good average results.

**Washing and Preserving Specimens**—After the sample is etched, it should be removed from the hot acid and washed under running water, and the "smut" deposited on the specimen should be removed by scrubbing with a stiff brush. Live steam is also an excellent method of washing etched work. The piece is first thoroughly rinsed and placed under the live steam nozzle. This results in rapid and complete drying with freedom from rust. For drying the specimen after washing, it can be blotted with a cloth or paper towel and dried with a blower.

For long preservation of the specimen, all traces of the acid should be removed either by rinsing the specimen in a caustic solution, in alcohol, or by boiling it several minutes in hot water. Then after the specimen has been dried it should be covered with a thin coat of transparent lacquer, or a film of oil which may be applied with the palm of the hand.

**Interpretation of Results**—The results obtained on properly etched samples are of great value if correctly interpreted. Surface seams, internal cracks, and pipes are easily recognized. It is the improper interpretation of the evidences of segregation and dendritic structures as revealed by deep etching that furnishes the greatest possibility of expensive errors and the needless rejection of material.

**Seams**—Seams in rolled material are of varying depth and usually extend in a straight path parallel to the direction of rolling. With forged material seams generally follow the contour of the forging and the flow of the metal. The most detrimental result of the presence of seams is their tendency to open up into deep hardening cracks if the metal is heat treated. This detrimental effect is avoided if, in the manufacturing operations, the surface is machined to sufficient depth to remove the seams.

**Surface Cracks**—Surface cracks as revealed by deep etching usually follow an irregular path and may result from improper handling during heating, forging or rolling, or during cooling from the finishing temperature. With heat treated material, surface cracks may be caused by improper treatment, by improper grinding after hardening, or by service stresses. It is worthy to note that surface cracks may be caused by macro etching hardened steel specimens that have not been properly tempered after hardening.

**Internal Cracks**—In addition to the macro etch, internal cracks and flakes can be detected by a fracture test on which they are revealed as brightly crystalline spots. The occurrence of internal cracks is detrimental to the steel and there is no means of overcoming their ill effect.

**Pipes**—Pipes are internal cavities and are invariably associated with segregated impurities which are deeply attacked by the etching reagent. Cavities in the center not associated with deeply attacked impurities are often mistaken for pipe, whereas such cavities can usually be traced to bursts caused from improper handling of the steel during forging or rolling.

**Segregations**—Segregations are revealed by the severity of the acid attack on the affected areas. The segregations may occur at the center



and be so deeply attacked that after etching they may appear as a pipe, or the segregation may be grouped in a rectangular form about the center. Nonmetallic inclusions also leave their mark by the presence of pits after macro etching.

Segregation as revealed by macro etching is not always an indication of defective material. The segregation revealed by macro etching should be examined under the microscope to determine if it is metallic segregation, crystalline arrangement, or a concentration of impurities.

**Dendritic Patterns**—Dendritic patterns are detected even in steel that has been subjected to repeated mechanical reduction. It has not been proven that dendritic patterns are associated with service failures, provided the material has been subjected to sufficient mechanical working and provided the segregation accompanying the dendritic formation is not in the form of nonmetallic inclusions which are incapable of dispersion.

In the formation of dendrite crystals, the intercrystalline material contains minute segregations which are largely diffused into the crystals. However, when nonmetallic inclusions, such as sulphides of manganese, are thus segregated, they are incapable of diffusion. The use of the microscope is suggested in such cases to supplement the macro examination.

## CONSTITUTION OF COPPER-PHOSPHORUS ALLOYS

By E. W. Roath\*

**P**RIOR to the work of Heyn and Bauer in 1907 the literature contains only results of simple chemical studies of the phosphides of copper. The earlier investigators found several phosphides, of which only the one with the formula  $\text{Cu}_3\text{P}$  is of commercial importance.

Several important experimental observations were made by Heyn and Bauer. Alloys higher than 15 per cent by weight of phosphorus cannot be prepared by fusion, but must be prepared by heating copper turnings with phosphorus at a lower temperature. Alloys richer in phosphorus are unstable at high temperatures and lose phosphorus. There tends to be a definite limiting value for each temperature. Thus at 1100 degrees Cent. the saturation content is 14.1 per cent phosphorus, which corresponds to the formula  $\text{Cu}_3\text{P}$ . In rapid heating and melting of phosphorus-rich alloys, as in casting into ingots, there is not sufficient time to reach the saturation limit of 14.1 per cent, and thus the commercial ingot of 15 per cent phosphorus is obtained.

The constitutional diagram in Fig. 1, except for the solid solubility line of phosphorus in copper, is from the data of Heyn and Bauer. They found that copper formed the eutectic with  $\text{Cu}_3\text{P}$  shown at C. The concentration of this eutectic is 8.27 per cent phosphorus and the melting point is 707 degrees Cent. The compound  $\text{Cu}_3\text{P}$  melts at 1022 degrees Cent., as do the alloys up to 15 per cent phosphorus. The alloys shown above 14.1 per cent phosphorus are mixed crystals of  $\text{Cu}_3\text{P}$ , and a second phosphide, probably  $\text{Cu}_5\text{P}_2$ . No solubility of copper in  $\text{Cu}_3\text{P}$  has been observed.

\*Research Dept., Chase Brass and Copper Co., Waterbury, Conn.

Heyn and Bauer placed the solid solubility of phosphorus in copper at 0.175 per cent on the basis of their microscopic observations of slowly cooled specimens. Hudson and Law, however, placed the limit at about 1 per cent

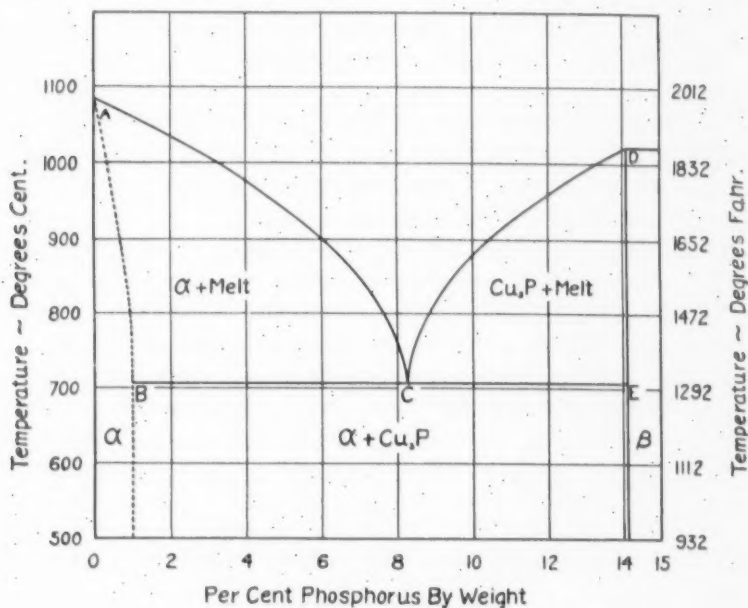


Fig. 1—Copper-Phosphorus Equilibrium Diagram.

phosphorus. They found that a phosphor copper containing 0.9 per cent of phosphorus by weight had an almost homogeneous structure after annealing for two hours at 690 degrees Cent., or four hours at 640 degrees Cent.

#### REFERENCES

- Schrötter, *Jahresberichte, Fortschritte der Chemie* (Liebig u. Kopp), 1849, p. 246.  
 Hvoslef, *Liebig's Annalen*, vol. C, p. 100.  
 Granger, *Comptes Rendus*, vol. CXIII, p. 1041; *Chemical News*, 1898, vol. 77, p. 227.  
 Maronneau, *Comptes Rendus*, vol. CXXVIII, 1899, p. 936, and 1900, vol. CXXX, p. 656.  
 Heyn and Bauer, *Metallurgie*, 1907, vol. IV, p. 242, and *Zeitschrift für Anorganische Chemie*, vol. 52, 1907, p. 129.  
 Guertler *Metallographie* (Gebrüder Borntraeger) vol. 1, 1912, p. 894.  
 Hudson and Law, *Journal, Institute of Metals*, 1910, vol. III, no. 1, p. 161.  
 Edwards and Murphy, *Journal, Institute of Metals*, vol. 27, 1922, p. 183.

## Comment and Discussion

### PRESERVATION OF FRACTURED SURFACES

**A**N interesting communication recently received from B. F. Shepherd, metallurgist, Ingersoll-Rand Company, Phillipsburg, N. J., described a method which he developed for the preservation of fractured surfaces of laboratory test specimens.

Mr. Shepherd says, "The problem of preserving fractures in the metallurgical laboratory has always caused us, at least, considerable difficulty. Attempts to coat with lacquer, etc., were not of much value. We recently found that wrapping a specimen in Cellophane or placing Cellophane over the fracture itself and cementing it to the steel by the following solvent makes an extremely effective preservative. The fractures can be seen through the Cellophane and they are entirely protected from the atmosphere or any acid fumes present. The solvent dissolves the Cellophane and makes the edges adhere firmly to the sides of the specimen. Solvent Formula: 18% Gum Arabic, 30% Glycerine, 51½% Water, ½% Formaldehyde.

"The Cellophane may be purchased from any of the concerns handling DuPont products and is fairly cheap."

### MEASURING MOLTEN METAL TEMPERATURES

**A**NOTHER communication received from John L. Ware of West Farmington, Ohio, comments on the use of the optical pyrometer for measurement of temperatures of metals in the ladle and during pouring.

It states that, "in order to more nearly approximate black body conditions in taking temperatures of molten metals in the ladle or during pouring it is found of considerable help to use a short refractory tube dipping into the metal and held with a rod or pair of long tongs. In actual practice an old muffle out of a laboratory electric furnace with the end knocked out answers well. For standardization an electric thermocouple pyrometer immersed in the metal (with suitable protection of course) may be read at the same time as the optical pyrometer sighted into the refractory tube partially immersed in the metal. This method has the advantage that results obtained are strictly comparative and may be corrected, if necessary to the true temperature."

### VOLUME XVI COMPLETED

**T**HIS issue of TRANSACTIONS completes Volume XVI, which covers the period from July, 1929, through December, 1929. The index for Volume XVI is now ready for distribution and may be secured upon request.

Those desiring to have their loose copies of Volume XVI bound in accordance with the style used in binding Volume I to XV, inclusive, may do so by forwarding them to the executive office of the society, 7016 Euclid Avenue, Cleveland, together with \$2.00 per volume, and they will be bound and returned promptly.

## Reviews of Recent Patents

By NELSON LITTELL, Patent Attorney  
475 Fifth Ave., New York City—Member of A. S. S. T.

**1,732,244, October 22, 1929, Method of Hardening Steel, Samuel I. Salzman, of Baltimore, Maryland.**

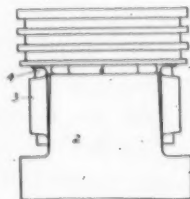
This patent describes a method and apparatus for hardening thin steel sheets in the form of razor blades to effect individual treatment of the blades without preventing a rapid rate of continuous manufacture. The



apparatus comprises a roll on which the strip of razor blades 10 is wound and drawn first through the furnace 18, then through the chilling members 20 and 21, consisting of water cooled chambers between which the strip is drawn, then through the channel frames 27 and 28 which are heated by the burners 29 and 30 and then through a second chilling member 31 and 32. The steel, when passing through the chilling members 20 and 21 and 31 and 32, is maintained free from contact of the chilling medium employed.

**1,732,557, October 22, 1929, Piston and Method of Making the Same, Zay Jeffries and Robert S. Archer, of Cleveland, Ohio, Assignors to Aluminum Company of America, of Pittsburgh, Pennsylvania, a corporation of Pennsylvania.**

This patent describes the production of a piston having a head 1, skirt 2, and bosses 3 attached to the head by a web 4, which piston is cast



from an aluminum base alloy containing between 6 and 16 per cent copper and 0.15 and 1.5 per cent of magnesium and heat treated by heating to a temperature slightly below the freezing point of the eutectic, quenching the casting and artificially aging by reheating at a temperature of about 150 degrees centigrade until the casting has attained a Brinell hardness number in excess of about 120.

**1,733,669, October 29, 1929, Method of Heat Treatment of Steel, Soji Makita, of Chatsu-Machi, Muroran-Shi, Hokkaido, Japan, Assignor to Kabushiki Kaisha Nihon Seikoshu, of Tokyo, Japan.**

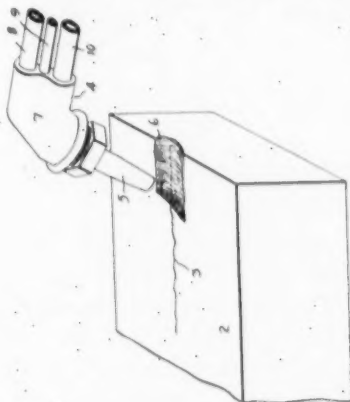
This patent describes a special heat treatment method for carbon and



alloy steels which gives superior qualities for all practical purposes as compared with the ordinary annealing methods. The purposes of the treatment are (1) to remove internal strains, (2) to soften the steel and make it more machinable, (3) to secure the desired strength and ductility and (4) to unify and impart a fine structure to the steel. The method consists in quenching the steel from a temperature above the transition point  $A_s$  then heating it up to a temperature a little higher than the point  $A_c$ , cooling it down to a temperature a little lower than the point  $A_r$ , heating it up to a temperature a little lower than the point  $A_c$ , maintaining that temperature for a period of time, and finally cooling it down.

**1,732,912, October 22, 1929, Method of Removing Defects from Steel Billets, Robert M. Rooke, Jr., of Jersey City, New Jersey, Assignor to Air Reduction Company, Incorporated, of New York, N. Y., a corporation of New York.**

This patent describes a method of chipping or removing seams, cracks, cold shuts and the like from billets by the use of an oxyacetylene



torch 4, the flame of which is directed from the nozzle at an inclination crosswise of the seam 3, so that the jet of oxygen is first directed obliquely downward and then as its velocity fails, is deflected upward again to burn out the seam, forming the broad curved depression 6.

**1,734,560, Nov. 5, 1929, Inhibitor, 1,734,561, Nov. 5, 1929, Inhibiting Compound, Ludwig J. Christmann, of Jersey City, New Jersey, Assignor to American Cyanamid Company, of New York, N. Y., a corporation of Maine.**

These patents describe the method of pickling and the use of an inhibitor to prevent the clean portions of the metal being attacked by the acid pickling solution. In the first patent the inhibitor consists of thio-benzamide and in the second of a condensation product of o-toluidine and sulphur chloride.

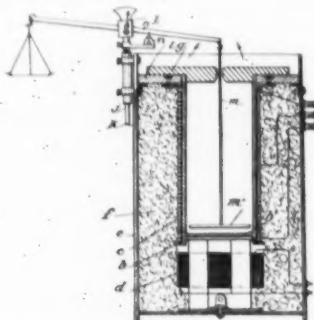
**1,734,928, Nov. 5, 1929, Treatment of Malleable Castings, Irving R. Valentine, of Erie, Pennsylvania, Assignor to General Electric Company, a corporation of New York.**

This patent describes a process for the treatment of malleable castings which are to be galvanized by hot-dipping in which the castings are

first cleaned by an acid pickling solution, then given a thin copper coating by dipping momentarily in a copper sulphate solution, then dipping in molten zinc and quenching the same, permitting the casting to age at a temperature between 120 and 145 degrees Cent. for a period of six hours.

**1,734,536, November 5, 1929, Electric Furnace, Victor Sorrel and Louis Andre Lafont, of Grenoble, France, Assignors to Ugine-Infra, of Grenoble, France.**

This patent describes an electric resistance furnace in which the temperature is automatically controlled by means of a balance 1 which suspends a magnetic body  $m'$  inside the furnace chamber and which is under



the influence of the electro-magnet d. As the temperature of the furnace is raised to the heating coils c, the magnetic properties of the body  $m'$  are reduced and the balance raises the member  $m'$ , opening the switch m to shut off the current to the coils c.

**1,734,930, Nov. 5, 1929, Temperature-Stabilized Shaft for Turbines and Method of Making the Same, Simon H. Weaver, of Schenectady, New York, Assignor to General Electric Company, a corporation of New York.**

This patent relates to the production of shafts for high speed turbines having a massive central portion 11 and reduced end portions 12. Elaborate directions are given for the production and heat treatment of these turbine shafts so as to produce a shaft which will not go out of alignment due to aging or other causes. The figure shows such a shaft supported on lathe centers 13 and 14 in a heat treating oven 15 heated by gas from the pipes 18. The baffle plate 19 prevents direct contact of the flame with the shaft.

**1,734,949, Nov. 5, 1929, Preservation of Metallic Surfaces, 1,734,950, Nov. 5, 1929, Means of Preserving Metallic Surfaces, James C. Vignos, of Nitro, West Virginia, Assignor to The Rubber Service Laboratories Co., of Akron, Ohio, a corporation of Ohio.**

These patents describe the process of pickling metal whereby the cleaning acid is prohibited from attacking the clean portions of the metal and causing pitting by the use of an inhibitor in the acid pickling bath. In the first patent the pickling acid is sulphuric acid and the inhibitor used is phenyl-p-tolyl-thiourea and in the second patent the inhibitor used is the reaction product of an aldehyde ammonia and thiocarbanilid. Directions are given in the patents for the preparation of the inhibitors.

# THE ENGINEERING INDEX

Registered United States, Great Britain and Canada.

In the preparation of the Engineering Index by the staff of the American Society of Mechanical Engineers some 1700 domestic and foreign technical publications received by the Engineering Societies Library (New York) are regularly searched for articles giving the results of the world's most recent engineering and scientific research, thought, and experience. From this wealth of material the A. S. S. T. is supplied with this selective index to those articles which deal particularly with steel treating and related subjects.

## AIRPLANE MATERIALS

**METALS**—Metals in Aircraft Construction, J. E. Hardecker. *Am. Mach.*, vol. 71, no. 15, Oct. 10, 1929, pp. 629-632, 7 figs.

Description of current factors and trends in use of ferrous and non-ferrous metals by airplane manufacturers; duralumin is outstanding metal for wing construction with riveting as means of fabrication; steel wires for bracing; use in fuselage of extruded duralumin angles and channels, fabricated by riveting; hull structures fabricated airplane engine construction.

## ALLOY STEEL

**SPECIFICATIONS**—Standard Specifications for Carbon Steel and Alloy Steel Blooms, Billets and Slabs for Forgings. *Am. Soc. Testing Matls. Standards—Supp.*, 1929, pp. 40-44.

Specifications cover: definition of term; scope; basis of purchase; process; discard; reduction from ingot; chemical composition; ladle analyses; check analyses; chipping; finish; marking; inspection; rejection; reheating.

## ALLOYS

Alloys (Legierungen), E. R. Thews. *Gießerei-Zeit. (Berlin)*, vol. 26, no. 16, Aug. 15, 1929, pp. 450-455.

First article of series giving review of special alloys compiled by author and arranged alphabetically; list includes over 3000 alloys which are described briefly; first instalment lists "A" aluminum alloy, aalener zinc alloy, Admiralty bronze, admos metal; adnic; advance metal; aerial; aero metal; aerolite; aeron; aich metal; aix leaf gold; ajax metals; aladar, alargan, albata, and albatra metal, etc. (To be continued.)

**AGE HARDENING**. Age Hardening. *Metallurgist (Supp. to Engineer, Lond.)*, Sept. 27, 1929, p. 130.

Remarks based on lengthy discussion of age hardening at meeting of Gesellschaft fuer Metallkunde at Dusseldorf; question upon which discussion centered was whether phenomena of age hardening could be accounted for by dispersion theory; certain number of German metallurgists held view

that phenomena are too complex to be accounted for by so simple an explanation; what may prove important step toward general acceptance of dispersion theory was explanation of way in which so-called anomalies in hardening phenomena could be readily fitted into dispersion theory.

**ANTI-CORROSIVE**—New Alloys Withstand Hydrochloric Acid, B. E. Field. *Chem. and Met. Eng.*, vol. 36, no. 9, Sept. 1929, p. 542.

Physical properties of new alloys made by Haynes Stellite Co., Kokomo, Ind., to meet special requirements in handling of acids and other corrosive chemicals; these are termed Hastelloy A, Hastelloy C, and Hastelloy D.

## ALUMINUM

Lighter Structures, R. L. Streeter. *Eng. News-Rec.*, vol. 103, no. 13, Sept. 26, 1929, pp. 487-491, 7 figs.

Two articles on important developments in aluminum industry. First article: what high-strength alloys of aluminum can do; average mechanical properties of strong aluminum alloy sheet; general survey of wrought alloys of aluminum in their various tempers; riveted and welded structural work of aluminum; casting alloys of aluminum; resistance to corrosion; aluminum roofs; aluminum and its strong alloys in industry; airplane gasoline tanks of aluminum. (To be concluded.)

## ALUMINUM ALLOYS

Aluminum and Its Alloys, A. G. C. Gwyer. *Engineer (Lond.)*, vol. 148, no. 3845, Sept. 20, 1929, pp. 294-295, 3 figs.

Review of development of heat treatment of aluminum alloys and application of methods of X-ray spectrography to aluminum.

Institute of Metals Autumn Lecture, A. G. C. Gwyer. *Metal Industry (Lond.)*, vol. 35, no. 11, Sept. 13, 1929, pp. 247-248, 1 fig.

Investigations of aluminum alloys are discussed; soundness and degasification; corrosion and protection; anodic oxidation. (To be continued.)

Aluminum and Its Alloys, A. G. C.

Those members who are making a practice of clipping items for filing in their own filing system may obtain extra copies of the Engineering Index pages gratis by addressing their request to the society headquarters, whereby their names will be placed on a mailing list to receive extra copies regularly.

Photostatic copies (white printing on a black background) of any of the articles listed may be secured through the A. S. S. T. The price of each print, up to 11 by 14 inches in size, is 25 cents. Remittances should accompany orders.

Gwyer. *Engineer (Lond.)*, vol. 148, no. 3844, Sept. 13, 1929, pp. 281-283, 8 figs.

Review of investigations carried out in recent years on aluminum alloys; advances made in knowledge of corrosion and protection; consideration of constitution of more important aluminum alloys, particularly aluminum-copper, aluminum copper-zinc, and aluminum silicon.

**HEAT TREATMENT.**—Electric Heat Treatment for Aluminum, N. B. Jones. *Metal Industry (N. Y.)*, vol. 27, no. 9, Sept. 1929, pp. 421-422, 4 figs.

Heat Treating aluminum alloy castings in cylindrical pit type electric furnace is described; theory of hardening aluminum alloys.

**PROTECTIVE COATINGS.**—Protective Varnishes for Aluminum Alloys. *Metalurgist (Supp. to Engineer)*, Aug. 30, 1929, pp. 125-126.

Article previously indexed from *Korrosion u. Metallschutz* Apr. 4, 1929.

#### ALUMINUM BRONZE

Aluminum Bronze Data. *Foundry*, vol. 57, no. 18, Sept. 15, 1929, supp. sheets, nos. 781 and 782.

Arranging so that some portion of casting which will not be wanted subsequently freezes last and takes consequent shrinkage; use of risers over danger points in casting aluminum bronze is discussed; devices insuring that riser is doing its duty as feeder up to last moment; restoring original properties of self-annealed castings for heating. (Continuation of serial.)

#### ALUMINUM CASTINGS

Aluminium-Kokillenguss, H. Obermueller. Berlin, V. D. I. Verlag, 1929, 84 pp., illus., diagrs., tables, 4 r.m.

Concise handbook on permanent mold casting and die casting; alloys used, construction of molds, casting processes, methods of testing, and equipment of plants are discussed from practical point of view. Eng. Soc. Lib., N. Y.

**X-RAY ANALYSIS.**—Radiography as a Tool in the Metal Industry, W. L. Fink and R. S. Archer. *Am. Soc. Steel Treating—Trans.*, vol. 16, no. 4, Oct. 1929, pp. 551-590 and (discussion) 590-599, 23 figs.

Typical examples are presented of applications made of radiographic method in laboratory of United States Aluminum Co.; cost of these radiographs is given and field of profitable applications of method is discussed. Bibliography.

#### AUTOMOBILE PARTS

**DEFECTS.**—Metallographic Studies of Defective Automobile Parts (Metallkundliche Untersuchungen von schadhafte Kraftfahrzeugteilen), M. v. Schwarz. *Zeit. des Bayerischen Revisions-Vereins (Munich)*, vol. 33, nos. 14, 15 and 17, July 31, Aug. 15 and Sept. 15, 1929, pp. 201-204, 219-221 and 246-249, 26 figs.

July 31: Analysis of fracture in front axle of automobile with four-wheel drive; broken transmission shaft; metallographic and chemical analysis of steering lever. Aug. 15 and Sept. 15: Injurious effect of grain-boundary or so-called crow-foot cementite

in ingot steel is demonstrated. (To be continued.)

#### CASE HARDENING

Progress in Case Hardening (Neuere Fortschritte auf dem Gebiete der Einsatzhaertung), A. Jaeschke. *Zeit. fuer die gesamte Giessereipraxis (Berlin)*, vol. 50, no. 34, Aug. 25, 1929, pp. 135-137.

It is claimed that advantages of calcium cyanide baths are greatly overrated; protective value of copper, nickel, and tin against cementation varies greatly; use of Schoop metal-spray process has proved uneconomic; accuracy of method of measuring depth of case; conditions which lead to failures in case hardening.

#### COPPER

**ROLLING.** Recrystallization of Electrolytic Copper After Hot Rolling (Die Rekristallisation von Elektrolytkupfer nach dem Warmwalzen), W. Tafel. *Zeit. fuer Metallkunde (Berlin)*, vol. 21, no. 8, Aug. 1929, pp. 265-267, 10 figs.

Comparative tests regarding different behavior of copper after hot rolling and iron after forging; for copper similar deviations from curves of Hanemann are observed as with soft iron; discussion of causes; test results and curves.

#### COPPER-TIN ALLOYS

The Eutectic Composition of Copper and Tin, G. O. Hiers and G. P. de Forest. *Am. Inst. Min. and Met. Engrs.—Tech. Pub.*, no. 241, Sept. 1929, 13 pp., 14 figs.

Experiments to determine eutectic composition of copper and tin, and location of part of liquidus line immediately above eutectic composition in copper; method of dissolving copper in bath of molten tin at constant temperature; differential cooling curves; microscopic examination; results of first method did not agree with results of Miller or of other investigators by other methods, nor with results of cooling-curve experiments and microscopic examination reported.

#### CRYSTALLIZATION

Crystalline Grains in Castings, A. Glazunov. *Foundry Trade J. (Lond.)*, vol. 41, no. 679, Aug. 22, 1929, pp. 131-134, 13 figs.

Lowest re-crystallization temperature shown in table for various metals; process of production of idiomorphic crystals; formation of nuclei from eutectic alloys; binary systems. Paper presented before Int. Foundrymen's Congress.

Crystalline Grains in Castings, A. Glazunov. *Foundry Trade J. (Lond.)*, vol. 41, no. 678, Aug. 15, 1929, pp. 117-120, 17 figs.

Microphotographic examination of crystalline structure of pure iron, pure lead, pure copper, galvanized iron sheet, brass and antimony castings; differences of opinion as to conception of crystallite are discussed; relation between crystal forming energy and surface tension is explained.

#### CRYSTALS

**STRUCTURE.** Determining Orientation of Crystals in Rolled Metal from X-ray



Patterns Taken by Monochromatic Pinhole Method. W. P. Davey, C. C. Nitchie and M. L. Fuller. *Am. Inst. Min. and Met. Engrs.—Tech. Publ.*, no. 243, Sept. 1929, 10 pp., 5 figs.

Outline of method; interpretation of patterns from one particular specimen of rolled zinc; results described are offered not in order to discuss effect of mechanical working on preferred orientations of zinc crystals, but merely as concrete example of application of method.

#### DIES, FORGING

The Construction of Drop Forging Dies. G. A. Smart. *Heat Treating and Forging*, vol. 15, no. 7, July 1929, pp. 841-842, 1 fig.

General considerations in selecting type of forging die to suit conditions in producing given forging are outlined; influence of hammer size on die design; considerations involved in small orders.

The Construction of Drop Forging Dies. G. A. Smart. *Heat Treating and Forging*, vol. 15, no. 8, Aug. 1929, pp. 1004-1006, 12 figs.

Special cases are given illustrating details that must have consideration in die design; distribution of flash and easy operation for hammerman.

The Construction of Drop Forging Dies. G. A. Smart. *Heat Treating and Forging*, vol. 15, no. 9, Sept. 1929, pp. 1169-1171, and 1175, 9 figs.

Importance of blockers in construction of drop-forging dies, and when they are to be used are discussed; ring-gear forging; sharp edges must be rounded; connecting rod forgings.

#### DIES, FORMING

Expansion Die for Forming Bead. Machy. (*Lond.*), vol. 34, no. 882, Sept. 5, 1929, pp. 613-714, 4 figs.

Design of expansion die used for forming bead around rim of cover made of  $\frac{1}{8}$ -in. sheet brass which is first drawn to shape in drawing die of standard design; die is so designed that circumference of expanded die has no opening or space between expanded sections.

#### FERRONICKEL CORROSION

The Corrosion Rate of Ferro-Nickel Alloys. C. G. Fink and C. M. DeCroly. *Am. Electrochem. Soc.—Advance Paper* no. 12, for mtg. Sept. 19-21, 1929, pp. 121-155, 15 figs.

Tests made on commercial alloys of iron-nickel series, using intermittent corrosion test; specifications of Am. Soc. for Testing Metals were followed closely; results in general indicate that rate of corrosion in sulphuric acid increased at first very rapidly with concentration of acid, reaches maximum then decreases again, reaching minimum, and usually increases gain to reach second maximum.

#### FORGINGS, STEEL

HEAT TREATMENT. How the Missouri Pacific Handles Alloy Steel. *Ry. Mech. Engr.*, vol. 103, no. 8, Aug. 1929, pp. 494-499, 12 figs.

Proper forging and heat treating methods

plus modern equipment result in maintenance economies and reduction of defects.

#### HARDNESS TESTING

A Synopsis of the Present State of Knowledge of the Hardness and Abrasion Testing of Metals. G. A. Hankins. *Mech. World (Lond.)*, vol. 86, no. 2221, July 26, 1929, pp. 78-80, 1 fig.

Suggestions for further research to be taken up by Hardness Tests Research Committee of Institution Mechanical Engineers; indentation hardness tests; Brinell test; ratio of tensile strength to Brinell number; diamond indentation tests; comparison of Brinell, Rockwell, and diamond pyramid scales. (To be continued.) Abstract of paper presented before Instn. Mech. Engrs.

A Synopsis of the Present State of Knowledge of the Hardness and Abrasion Testing of Metals. G. A. Hankins. *Mech. World*, vol. 86, no. 224, Aug. 16, 1929, pp. 146-147, 2 figs.

Results of investigation of direct abrasion by sand, carried out by Brinell and reviewed by Holz; discussion of diamond scratch test and comparisons of investigations by author with those of Hadfield and Main; Jertz theory of hardness; magnetic hardness test. (Continuation of serial.)

Coercive Force and Mechanical Hardness (Koerzitivkraft und mechanische Haerte). A. Kussmann and R. Scharnow. *Zeit. fuer Physik (Berlin)*, vol. 54, no. 1-2, March 21, 1929, pp. 1-15, 6 figs.

Study of relationship between coercive force (magnetic hardness) and mechanical hardness of alloys and their dependence upon structure of alloys; coercive force is shown to exhibit strong increase in case of heterogeneous mixtures where it is independent of hardness; experimental and theoretical details.

#### HARDNESS TESTING MACHINES

Methods of Hardness Testing. *Am. Mach.*, vol. 71, no. 7, Aug. 15, 1929, p. 297, 1 fig.

Monotron described is static mechanical-pressure machine employing small diamond-ball impresser point for measuring hardness of metals, minerals; glass, porcelain, brick, wood, rubber, fiber, paper, and other materials; two dials are used to take readings, one measuring pressure applied and other depth of impression under load; pressure scale reads up to 160 kg., or 352 lb., and micrometer depth gage reads in 1/5000 in. and also in 1/200 mm.

#### INGOT MOLDS

New Low-Sulphur Iron for Ingot Molds. *Iron Age*, vol. 124, no. 5, Aug. 1, 1929, p. 280, 2 figs.; see also *Iron Trade Rev.*, vol. 85, no. 5, Aug. 1, 1929, p. 260, 2 figs.

Description of ingot mold iron developed by Vulcan Mold and Iron Co., Latrobe, Pa., to resist as much as possible failure of molds through heat checking and cracking; 25 per cent longer service claimed for ingot molds made of new low sulphur iron; size and distribution of graphite controlled.

#### IRON ALLOYS

ANALYSIS. Recent Developments in the Analysis of Carbon in Iron and Iron Alloys. N. A. Zeigler. *Am. Electrochem.*

*Soc.—Advance Paper*, no. 1, for mtg. Sept. 19-21, 1929, pp. 1-8, 3 figs.

Modifications and improvements in "Yensen" method for accurate determination of carbon in iron and iron alloys; error in method is approximately 0.01 mg., corresponding to approximately 0.0005 per cent for 2-gram sample; with extra precautions error can be reduced to approximately 0.0001 per cent.

**MAGNETIC PROPERTIES.** On the Intensity of Magnetization in Iron-Nickel-Cobalt Alloys, H. Masumoto, *Tohoku Imperial—Science Reports (Sendai, Japan)*, vol. 18, no. 2, June 1929, pp. 195-229, 10 figs.

Intensity of magnetization of binary and ternary alloys of three ferromagnetic substances, iron, nickel, and cobalt, was measured by ballastic method, and relation between magnetization and equilibrium diagrams of these systems investigated; it was found that magnetization curve of reversible alloys has general form similar to that of pure iron or nickel.

**PROPERTIES.** Recent Research on the Physical Properties of Iron Alloys (Neuere Untersuchungen ueber die physikalischen Eigenschaften der Eisenlegierungen), A. Schulze, *Gieserei-Zeitung (Berlin)*, vol. 26, nos. 14 and 15, July 15 and Aug. 1, 1929, pp. 389-398 and 428-434, 37 figs.

Review of investigations of most important physical properties of iron-carbon, iron-nickel, iron-silicon, iron-aluminum, iron-manganese, iron-cobalt, iron-chromium, and stainless steels; influence of different alloying elements on behavior of iron; comparison of properties of iron alloys.

X-Ray Analysis. X-Ray Investigation of Iron and Zinc Alloys, A. Osawa and Y. Ogawa, *Tohoku Imperial Univ.—Science Reports (Sendai, Japan)*, vol. 18, no. 2, June 1929, pp. 165-176, 5 figs.

The result of X-ray analysis confirmed correctness of new diagram of Fe-Zn system recently published by one of writers.

#### IRON-CHROMIUM ALLOYS

**MICROSTRUCTURE.** Study of the Iron-Chromium-Carbon Constitutional Diagram, V. N. Krivobok and M. A. Grossmann, *Am. Soc. Steel Treating—Preprint for mtg.*, Sept. 9-13, 1929, no. 18, 36 pp., 53 figs.

Changes in micro-structure of iron-chromium-carbon alloys, caused by variation in chromium content with various carbon contents; previous work; alloys examined, up to chromium content of 35 per cent and carbon content of 0.60 per cent; region of mixed structures (alpha delta iron plus austenite) is extensive feature in range from 20 to 35 per cent chromium; region of pure austenite greatly reduced in extent; section of ternary constitutional diagram constructed.

#### IRON-SILICON ALLOYS

Iron-Silicon-Carbon Alloys. Constitutional Diagrams and Magnetic Properties, *Iron and Steel Inst.—Advance Paper*, no. 12, Sept. 1929, 17 pp., 7 figs.

Author reviews and discusses work done in regard to magnetic properties and metallography of iron-silicon-carbon alloys; based on these results, hypothesis is presented that

alpha iron, even in absence of silicon, can exist at all temperatures below freezing point, and that gamma modification is due to carbon and other interstitial impurities and consequently that gamma iron is not inherent characteristic of iron.

#### IRON AND STEEL

**PHOSPHORUS DETERMINATION.** Phosphorus Determination, Iron and Steel, R. P. Hudson, *Heat Treating and Forging*, vol. 15, no. 8, Aug. 1929, pp. 995-998.

Detailed outline of following methods: molybdate-magnesia method, weighing yellow precipitate, Emmerton's volumetric method, alkalimetric method, preparation and standardization of potassium permanganate solution.

#### IRON AND STEEL PLANTS

**METALLURGY.** Comments on Cleanliness of Steel, G. W. Walker and R. E. Sherlock, *Heat Treating and Forging*, vol. 15, no. 9, Sept. 1929, p. 1176.

Discussion of paper presented before American Iron and Steel Institute which was published in June issue of magazine; to say that method of inspection for cleanliness should be based on average size, average amount and uniformity of distribution of non-metallics, as observed in reasonable number of micro-sections is to entirely misconceive purpose of such inspection.

#### MAGNESIUM

High Purity Magnesium Produced by Sublimation, H. E. Bakken, *Chem. and Met. Eng.*, vol. 36, no. 6, June 1929, pp. 345-347, 4 figs.

Description of process developed to purify metal made for relatively impure and cheap oil; process is purification by distillation with arrangement of retort and condenser to produce metal vapor which will condense to liquid of high purity.

#### MAGNESIUM ALLOYS

Ultra-light Alloys, *Machy. (Lond.)*, vol. 34, no. 873, July 4, 1929, pp. 434-435.

Properties and applications of magnesium alloys are described; they allow greatest possibilities in production of alloys; density is only 1.738; good endurance properties; tendency to corrode readily and low limit of proportionality are to some extent overcome by suitable alloying and heat treatment; properties of Dow metal alloys given.

Aluminum, die Leichtmetalle und ihre Legierungen, P. Molchior; im Auftrage der Deutschen Gesellschaft fuer Metallkunde, Berlin, V. D. I. Verlag, 1929, 280 pp., illus., tables, diagrs., 15-r.m.

Handbook on aluminum and magnesium and their alloys, which aims particularly to meet practical needs of engineers; metallography, chemistry and physical properties of metals are described; methods of shaping, finishing and joining are explained; uses of metals and alloys for various purposes are related; standards of various countries are given; much scattered scientific information is made conveniently accessible to machine builders, *Eng. Soc. Lib., N. Y.*

#### MANGANESE STEEL

Austenite Manganese Steel, J. H. Hall.

*Am. Soc. for Steel Treating*—TRANS., vol. 16, no. 3, Sept. 1929, pp. 458-460.  
Composition, method of manufacture, heat treatment, properties and hardening power are briefly described.

### MATERIALS TESTING MACHINES

Adjusting of Metal Testing Machinery (Les procédés de tarage des machines pour l'essai des métaux). *Revue Industrielle (Paris)*, vol. 59, no. 2241, Aug. 1929, pp. 563-570, 10 figs.

Indirect and direct methods of calibration by means of weight, transit bars, crushers, steel bands below electric limit, piezometric rings, etc.

### METALLOGRAPHY

Metallography Simplified for Practical Use in Shop. E. Preuss, G. Berndt and M. v. Schwarz. *Iron Trade Rev.*, vol. 85, no. 4, July 25, 1929, pp. 196-198, 4 figs.

Discussion of microscopic testing by other metallographic methods; distinguishing between different types of steel by spark test; special apparatus for determination of critical points; double-curve recorder of Siemens and Halske; critical point apparatus manufactured by Leeds and Northrup Co. (To be concluded.)

Methods of Research in Metallography. G. Masing. *Inst. of Metals—Advance Paper (Lond.)*, no. 507, for mtg. Sept. 9-12, 1929, 19 pp., 10 figs.

Fundamental principles underlying methods of research employed in investigations on constitution of alloys and in physical metallography are discussed, and it is shown that researches on constitution of alloys must be based on well-established thermodynamical laws of heterogeneous equilibria; experimental methods are, however, not sufficient for accurate work and need to be improved precision and amplified; most important improvement required is development of method for rapid attainment of equilibrium in alloy.

### METALS

COLD WORKING. On the Distribution of Hardness Produced by Cold Working. W. P. Sykes and A. C. Ellsworth. *Am. Soc. Steel Treating—Preprint*, for mtg. Sept. 9-13, 1929, no. 3, 11 pp., 6 figs.

Preliminary study of progress of hardening throughout section produced by cold working; copper, Armco ingot iron, and carbon free alloy of iron with 8 per cent tungsten, differing considerably in hardness, are reduced by cold swaging; hardness measurements taken at various depths below surface of round section after each series of reductions; effects of starting diameter and original hardness upon distribution of final hardness.

CORROSION PREVENTION. Modern Trend in Fighting Corrosion. T. H. Turner. *Heat Treating and Forging*, vol. 15, no. 7, July 1929, pp. 860-864.

Outstanding success achieved by development of thick deposits of nickel, with possibly composite layers of copper and chromium in electroplating, by production of martensitic and austenitic stainless and super-stainless steels in ferrous metallurgy, by replacement of brass by cupro-nickel and

monel metal in non-ferrous metallurgy, and introduction of manganese-nickel and chromium-nickel alloys and heat-resistant steel for use at high temperatures. Bibliography. Abstract of paper before Sheffield Metallurgical Assn.

CORROSION RESISTANCE. Testing Method for Judging Corrosion Resistance of Metals Against Weather and Seawater (Prüfverfahren zur Beurteilung der Korrosions-Bestaendigkeit von Metallen gegen Witterung und Seewasser). E. Rackwitz and E. K. O. Schmidt. *Zeit. fuer Flugtechnik und Motorluftschiffahrt (Munich)*, vol. 20, no. 6, Mar. 28, 1929, pp. 137-141, 10 figs.

Series of methods developed for light metals, by DVL (German Testing Institute for Aviation); results show that by further development of equipment tests will be possible, answering practical conditions in wide range; these series of tests are not considered to form complete whole yet.

DEFORMATION. Discontinuous Rate of Deformation (Ueber die sprungartige Deformation). M. Classen-Nekudowa. *Zeit. fuer Physik (Berlin)*, vol. 55, no. 7-8, June 20, 1929, pp. 555-568, 13 figs.

Report on studies made at Leningrad X-Ray Institute of Engineering Physics, following pioneer work of A. Joffe; plastic deformation of salt, brass, and aluminum single crystals and groups of crystals at high temperatures occurs in jumps, magnitude of which is function of temperature.

ELECTRIC CONDUCTIVITY. The Theory of Electrical Conductivity. W. V. Houston. *Am. Inst. Elec. Engrs.—Advance Paper*, no. 126, for mtg. Sept. 3-6, 1929, 5 pp., 3 figs.

Study of behavior of electrons in metals, and discussion of various theories of electric properties of metals.

PROPERTIES. The Problem of the Metallic State. J. D. Bernal. *Faraday Soc.—Trans.*, vol. 25, no. 98, July 1929, pp. 367-379, 1 fig.

Characteristics of true metals are considered together with those of compounds and solid solutions and hypothesis as to nature of superconductivity is put forward; importance of magnetic measurements is emphasized and explanation in terms of homopolar binding is given to certain diamagnetic metallic compounds which follow Hume-Rothery rules; empirical requirements of metallic bond are discussed with their possible theoretical meaning.

### METALS CORROSION

The Corrosion of Metals. W. Palmer. *Ingeniörs Vetenskaps Akademien*, no. 93, 1929, 347 pp., 101 figs.

Historical introduction, theory of local galvanic elements; experimental researches for testing of theories; choice between three theories discussed.

The Prevention of Corrosion. E. S. Stokes. *Chem. Eng. and Min. Rev. (Melbourne)*, vol. 21, no. 251, Aug. 5, 1929, pp. 432-440.

Review of methods adopted or suggested for prevention of corrosion; increasing resistance of metal; controlling or eliminating corrosive influences; applying protective coating; discussion of pH scale.

Atmospheric Corrosion of Metals. J. C.



Hudson. *Faraday Soc.—Trans. (Edinburgh)*, vol. 25, no. 100, Sept. 1929, pp. 475-496, 4 figs.

Results of test showing that corrosion of specimens is directly proportioned to pollution of atmosphere; dezincification of brass by atmospheric corrosion was also observed; resistance of specimens undergoing continuous test.

#### METALS FATIGUE

High Frequency Fatigue, C. F. Jenkin and G. D. Lehmann. *Roy. Soc.—Proc. (Lond.)*, vol. 125, no. A796, Aug. 1, 1929, pp. 83-117 and (discussion) 117-119, 24 figs.

Researches to determine effect of frequency of alternation of stress on fatigue limits of various metals were carried out at Engineering Laboratories, Oxford, tested rolled, normalized and hardened steel, rolled aluminum, annealed copper and normalized armco iron; fluctuations of air pressure acting directly on specimen were used to make them vibrate; increase of fatigue limit up to 60 per cent recorded.

The Fatigue Testing of Sheet Metals, J. R. Townsend and C. H. Greenwall. *Metal Stampings*, vol. 2, no. 8, Aug. 1929, pp. 599-600.

Description of testing procedure; results of cold work on physical properties of non-ferrous metals as indicated by fatigue tests.

The Influence of Oxygen on Corrosion Fatigue, A. M. Binnie. *Engineering (Lond.)*, vol. 128, no. 3318, Aug. 16, 1929, p. 190, 1 fig.

Experiments described were carried out in Engineering Laboratory at Oxford to determine to what extent reduction of fatigue limit of certain steels by corrosion is due to pressure of oxygen in surrounding atmosphere; specimens used were cut from same bar and were tested in Woehler rotating-cantilever machine at 2000 r.p.m.; fatigue limit in air, at ten million reversals, is 17.0 tons per sq. in., this being calculated from nominal bending stress.

The Correlation of Fatigue and Overstress, J. H. Smith and F. H. Armstrong. *Iron and Steel Inst.—Advance Paper (Lond.)*, no. 10, Sept. 1929, 29 pp., 26 figs.

Brief summary of recent methods used in attempt to give satisfactory practical explanation of phenomenon of fatigue of metals; strain method was used in tests to determine variation of fatigue limits resulting from overstress; yield ranges of steel which has failed in practice were investigated; description of testing machine; test methods and results.

#### METALS TESTING

Press Working and Forming of Metals, E. V. Crane. *Metal Stampings*, vol. 2, no. 8, Aug. 1929, pp. 585-590, 5 figs.

Discussion of methods of determining rate of strain hardening of metals; effect of direction of loading on nature of deformation.

Testing of Metals and Alloys at High Temperatures for Compression and Drawing (Les essais a chaud des metaux et alliages par compression et par filage), A. Portevin and F. Le Chatelier. *Academie des Sciences—Comptes Rendus (Paris)*, no. 5, July 29, 1929, pp. 248-250.

Discussion of experimental data and derivation of empirical formula.

Solubility of Gas in Melting of Pure Aluminum and Aluminum Alloy (Gasloeslichkeit in Schmelzen von Reinaluminium und einer Aluminiumlegierung), S. Briesemeister. *Zeit. fuer Metallkunde (Berlin)*, vol. 21, no. 8, Aug. 1929, pp. 268-270, 7 figs.

Aluminum and alloy consisting of 8 per cent copper and 92 per cent aluminum are investigated under gasification with nitrogen, oxygen, carbon monoxide, carbon dioxide, sulphur dioxide, hydrogen, illuminating gas, methane, ethylene and water vapor; remarks regarding "Y" alloy.

BIMETALS. Bimetal (Bimetall), W. Rohn. *Zeit. fuer Metallkunde (Berlin)*, vol. 21, no. 8, Aug. 1929, pp. 259-264, 11 figs.

Bimetal is determined as two strips of metal of different thermal stretch and which bend under temperature influences; performance of bimetals; thermal stretch of metals and alloys; considerations regarding suitable bimetal combinations; testing equipment for bimetals; technological viewpoints in fabrication and working of bimetals.

#### MICROSCOPES

Preparing Thin Specimens for Microscopic Examination, R. A. Ragatz. *Min. and Met.*, vol. 10, no. 272, Aug. 1929, pp. 372-379, 18 figs.

Various operations in preparation of specimen for microscopic study offer interesting problems in technique in case of thin specimens.

#### MOLYBDENUM STEEL

On the Equilibrium Diagram of the Iron Molybdenum System, T. Takei and T. Murakami. *Am. Soc. for Steel Treating—TRANS.*, vol. 16, no. 3, Sept. 1929, pp. 339-358 and (discussion) 358-371, 56 figs.

Equilibrium diagram of iron-molybdenum system studied by microscopic investigation, electric resistance method, and by dilatometric and magnetic analyses; in this system, two intermetallic phases, epsilon and eta, exist; solubility of molybdenum in alpha iron; in alloys containing more than 63 per cent of molybdenum, eutectoid is found, consisting of eta-phase and molybdenum dissolving iron. See Engineering Index, 1928, p. 1194.

#### NICKEL-CHROMIUM IRON

BRITTLENESS. The Brittle Range In 18 And 8 Chromium-Nickel Iron, H. H. Lester. *Am. Soc. Steel Treating—Preprint for mtg.* Sept. 9-13, 1929, no. 10, 20 figs.

Investigation of chromium-nickel iron containing 18 per cent chromium and 8 per cent nickel, made by Army Ordnance Dept.; critical point determination; series of tensile tests; microscopical examination; iron austenitic at room temperature and relatively brittle on heating to about 1300 deg. Fahr.; well defined critical point at 1330 deg. Fahr. on heating and at 1148 deg. Fahr. on cooling; loss of ductility due to formation of ferrite crystals and to precipitation of iron or chromium carbide along atomic planes in austenite grains.

#### NITRIDATION

Notable Symposium on Nitriding. *Iron*



g of Pure Alu-  
Gasloeslichkeit  
ium und einer  
semeister. Zeit.  
vol. 21, no. 8,  
s.  
sting of 8 per  
aluminum are  
n with nitro-  
e, carbon diox-  
n, illuminating  
water vapor;

Bimetall), W.  
nde (Berlin),  
p. 259-264, 11

two strips of  
stretch and  
e influences;  
mal stretch of  
ions, regarding  
testing equip-  
ical viewpoints  
f bimetal.

s for Micro-  
Ragatz. Min.  
Aug. 1929, pp.

ration of speci-  
ffer interesting  
of thin speci-

m of the Iron  
Fakai and T.  
eel Treating—  
1929, pp. 339-  
56 figs.

on-molybdenum  
e investigation,  
and by dilato-  
s; in this sys-  
s, epsilon and  
denum in alpha  
re than 63 per  
oid is found,  
olybdenum dis-  
g Index, 1928,

N  
ittle Range In  
Iron, H. H.  
ating—Preprint  
10, 20 figs.

m-nickel iron  
ium and 8 per  
rdnance Dept.;  
series of ten-  
mination; iron  
e and relatively  
300 deg. Fahr.;  
1330 deg. Fahr.  
Fahr. on cool-  
o formation of  
itation of iron  
atomic planes in

Nitriding. Iron

Age, vol. 124, no. 12, Sept. 19, 1929, pp. 738-741, 3 figs.

Review of symposium on nitriding at National Metal Congress in Cleveland with abstracts of papers presented; latest developments described by A. Fry; high-frequency furnace and advantages of older processes discussed by C. K. Everitt; Few Practical Fundamentals of Nitriding Process, H. W. McQuaid; Investigations in Nitriding, R. Sergeson; Short-Time Nitriding of Steels in Molten Cyanides, A. B. Kinzel and J. J. Eagen.

Hardening Steel in Ammonia Gas. *Iron Trade Rev.*, vol. 85, no. 10, Sept. 5, 1929, pp. 597-598, 1 fig.

Description of nitriding equipment and operations employed by American Metal Treatment Co., Elizabeth, N. J., subsidiary of American Gas Furnace Co.; experiments conducted by American Gas Furnace Co., indicate better results with relatively low gas concentration and greater time element.

Surface Hardening of Steel by Nitrogen, H. W. McQuaid. *Mech. Eng.*, vol. 51, no. 9, Sept. 1929, pp. 667-671, 4 figs.

Discussion of position of nitrided steels as engineering materials; characteristics of nitrided steel; explanation of process; results obtained with process by Timken-Detroit Axle Co.

## NONFERROUS ALLOYS

Nonferrous Metals and Alloys. *Soc. Automotive Engrs. Jl.*, vol. 25, no. 3, Sept. 1929, pp. 305-306.

Information on non-ferrous metals and alloys which is not ready for incorporation in specification form is given by Nonferrous Metals Division of Standards Committee; properties and applications of non-ferrous cutting tools and die materials described including; Stellite, cemented tungsten carbide, and Elkonite; tungsten and platinum used in automotive ignition systems and incandescent lamps.

CORROSION. Tarnishing the Nonferrous Metals and Alloys, B. Jeffs. *Black and White*, vol. 2, no. 2, Sept. 1929, pp. 9-10.

Study of corrosion of different metals shows that products of corrosion may be divided into following classes which are discussed; films which accelerate corrosion, films which neither accelerate nor retard corrosion, but allow corrosion to proceed at uniform rate, and films which retard corrosion.

PROPERTIES. Engineering of Nonferrous Metals, J. R. Townsend. *Iron Age*, vol. 124, no. 8, Aug. 22, 1929, pp. 457-458.

Essential requirement of all materials of construction are reviewed; economic factors governing use of special alloys; failures which may occur after stresses are resisted; rigid specifications are necessary for mass production.

VOLATILITY. Volatility of Heavy Metal Compounds in Metallurgical Process (Zur Flüchtigkeit von Schwermetallverbindungen in huettenmaennischen Prozessen), Kohlmeier. *Metall und Erz (Berlin)*, vol. 26, no. 3, Feb. 1, 1929, pp. 62-66, 2 figs.

Compilation of information on vapor tension and volatility of most important metals, their oxides and sulphides; new determinations of boiling and sublimation points of

lead oxide and cadmium oxide; volatilization of tin and lead in bessemerization of copper ores. Report from department of metallurgy of Berlin Inst. of Technology.

VOLATILITY. Volatility of the Oxides of Lead, Cadmium, Zinc and Tin (Ueber die Flüchtigkeit der Oxyde von Blei, Cadmium, Zink und Zinn), Feiser. *Metall und Erz (Berlin)*, vol. 26, no. 11, June 1929, pp. 269-284, 14 figs.

Abridged doctor dissertation of department of metallurgy of Berlin Institute of Technology on determination of vapor tension and sublimation pressure of oxides of lead and cadmium oxides to 1200 deg., of oxides of zinc and tin up to 1500 deg., by special method based on loss of weight; determination of boiling point of oxides of lead and cadmium; effect of oxides of very low volatility, on volatility of lead oxide; analysis of converter dust.

## PICKLING

Pickling Methods, W. G. Imhoff. *Black and White*, vol. 2, no. 2, Sept. 1929, pp. 11-14, 3 figs.

Discussion of test made to determine results of pickling and influence of different methods upon quality of product.

## RAILS

MANGANESE STEEL. Medium Manganese Steel Makes Higher Grade Rails, C. B. Bronson. *Ry. Age*, vol. 67, no. 9, Sept. 1929, pp. 563-565, 12 figs.

Revival of use of material that formerly gave excellent service, results in longer life and greater safety; Bessemer rail proves unsatisfactory; advantages in manufacture and use; diagrams showing wear of various steel rails.

TESTING. Effect of Service on the Endurance Properties of Rail Steel, J. R. Freeman, Jr. and H. N. Solakian. *U. S. Bur. of Standards—Jl. of Research*, vol. 3, no. 2, Aug. 1929, pp. 205-246, 35 figs.

Series of tests have been carried out to determine effect of service on endurance properties of rail steels; data are given on endurance properties of steel from rails from same heats after service and from two rails which failed due to transverse fissure; it was found that endurance properties of steel were not affected by over 20,000,000 tons of traffic.

TESTING. Progress on the Problem of Transverse Fissures in Rails, M. E. McDonnell. *Metals and Alloys*, vol. 1, no. 3, Sept. 1929, pp. 93-97, 3 figs.

Discussion of various investigations and testing methods. Bibliography.

## SHEET METAL

BLISTERING. Conditions Necessary for Blistering of Metal During Processing, A. Hayes. *Am. Soc. Steel Treating—Preprint for mtg.* Sept. 9-13, 1929, no. 14, 10 pp., 3 figs.

Two facts not generally recognized in considering causes of blistering of sheet metal are pointed out; pressure required to raise blister inversely proportional to radius of curvature of lamination in sheet; sources of gases which might produce blister formation; pressures responsible for penetration of hydrogen during pickling process due to sur-

face-tension of pickle liquor during time that hydrogen bubbles are quite small.

#### SHEET STEEL

**HEAT TREATMENT.** Continuous Heat Treatment of Sheet Metal, E. S. Lawrence. *Heat Treating and Forging*, vol. 15, no. 9, Sept. 1929, pp. 1160-1163 and 1168, 7 figs.

Process of normalizing, its beneficial results and general design of furnaces for its accomplishment are outlined; continuous heat-treating equipment requisites; conveyor mechanisms in continuous furnace; furnace dimensions increase; temperature ranges; fine grain structures produced are shown.

#### SPRING STEEL

**HEAT TREATMENT.** Driving Wheel Springs and Equipment for Spring Repairs. *Ry. Mech. Engr.*, vol. 103, no. 8, Aug. 1929; pp. 483-489, 8 figs.

In manufacture and repair of railroad leaf-springs two important features are involved; through understanding of four fundamental requirements of heating and treating to obtain proper physical characteristics, and combining with above practical shop details entering performance of work; operating heating furnace for forming and quenching as well as furnace for drawing back; process explained; practice for manufacture and repairs of elliptic springs.

#### STAINLESS IRON

Stainless Iron and Steel, A. G. Lambert. *Can. Machy. and Mfg. News (Toronto)*, vol. 40, no. 17, Aug. 22, 1929, pp. 60-61, 1 fig.

Analysis and properties of principal types of stainless iron and steel in commercial use; brief indication of wide applications as structural work, turbine blades, pump rods, machine parts, chains and conveyers, golf clubs, etc.

#### STAINLESS STEEL

Stainless Iron and Steel, A. G. Lambert. *Can. Machy. and Mfg. News (Toronto)*, vol. 40, no. 17, Aug. 22, 1929, pp. 60-61, 1 fig.

Analysis and properties of principal types of stainless iron and steel in commercial use; brief indication of wide applications as structural work, turbine blades, pump rods, machine parts, chains and conveyers, golf clubs, etc.

**AUTOMOBILE.** Stainless Steel's Advantage Lies in Non-Corrosive Properties, A. F. Denham. *Automotive Industries*, vol. 61, no. 10, Sept. 7, 1929, pp. 341-342, 1 fig.

Possibilities of stainless steel for bright metallic exterior parts of automobiles are discussed; comparison made with other types of metallic exterior finishes; nickel plating as final finish; chrome plating over copper, nickel, with steel base; chrome plating over brass; nickel and chrome alloy stainless steels; stainless steel with non-metallic inclusion, comprising zirconium sulphide or molybdenum sulphide; cadmium plating.

#### STEEL ANALYSIS

**CARBON CONTENT.** Calorimetric Carbon Test Improved, H. L. Campbell. *Iron Age*, vol. 124, no. 3, July 18, 1929, pp. 157-158, 2 figs.

Description of calorimetric method of an-

alyzing steel for carbon; equipment required consists of analytical-balance, water bath, calorimeter with glass tubes, and test tubes.

**CARBON SOLUBILITY.** The Solubility of Carbide in Ferrite, H. A. Dickie. *Iron and Steel Inst.—Advance Paper (Lond.)*, no. 2, Sept. 1929, 25 pp.

Description of investigation to determine effect of various elements on solubility of carbide in ferrite and to determine form of solubility curves in wide range of steels; changes due to tempering effects, such as spheroidization, have been eliminated from results by methods of heat treatment adopted, and solubility curves for 18 steels used have been determined and reduced to terms of carbon content.

#### STEEL CASTINGS

**COPPER MOLDS.** Casting Steel in Water-Cooled Copper Moulds, W. Oertel. *Iron and Coal Trades Rev. (Lond.)*, vol. 118, no. 3198, June 14, 1929, pp. 898-899.

Mold suitable for ingots weighing 0.2 metric ton; melts made in Heroult furnace; surface of ingots smooth, and without pores or cracks; tabular data on chemical composition of melts and on results of tensile tests, hardness, and magnetic properties. Abstract translated from Stahl u. Eisen.

**HEAT TREATMENT.** Heat-Treatment of Steel Casting, "Acero." *Iron and Steel Industry (Lond.)*, vol. 2, no. 12, Sept. 1929, pp. 399-400.

Essential requirements for heat treatment of steel castings are discussed covering annealing, normalizing, quenching, and tempering; results of tests of hard-steel and mild-steel castings.

**MICROSTRUCTURES.** Various Conference Papers Criticised, G. F. Gillott, H. Van Aarst, Diepschlag, and K. Giederiejowski. *Foundry Trade J. (Lond.)*, vol. 41, no. 680, Aug. 29, 1929, p. 156.

Discussion following presentation of four papers before Inst. Brit. Foundrymen, is given covering critical examination of steel castings, problem of foundry molding sands, influence of graphitization of cast iron, and elimination of sulphur from alloys of iron.

**TEMPERATURE EFFECT.** Mechanical Properties of Cast Steel at High Temperatures (Mechanische Eigenschaften von Stahlguss bei erhöhten Temperaturen), F. Koerber and A. Pomp. *Zeit. des Bayerischen Revisions-Vereins (Munich)*, vol. 33, no. 13 and 14, July 15 and 31, 1929, pp. 195-196 and 211-212, 4 figs.

Results of tests carried out at temperatures up to 500 deg. cent. on 12 different cast steels. Abstract from Mitteilungen des Kaiser-Wilhelm Inst. fuer Eisen-Forschung.

#### STEEL FORGINGS

Difficulties Encountered in Forging Steels, H. W. McQuaid. *Heat Treating and Forging*, vol. 15, no. 9, Sept. 1929, pp. 1166-1168.

Review of factors of steel quality in connection with manufacture of quality forgings that will fully satisfy user; causes of difficulties are described including forging laps, burnt forgings, piped steel, seamy stock, and inability to secure heat treatment results.

**STEEL HARDENING**

Salt Baths for the Hardening of High Speed Steel, J. Kjerrman. *Am. Soc. for Steel Treating—Trans.*, vol. 16, no. 3, Sept. 1929, pp. 393-404, 2 figs.

Results of series of experiments on salt baths containers used for heating high-speed steel for hardening; decarburization of metal being heated is one of difficulties encountered; effect of crucible material, salt used and combinations of two are reported; chamotte crucible as container for barium chloride, to which sufficient amount of ferrosilicon was added, could be used.

**TENSILE.** Tension Tests of Steel at High Temperature (Essais de traction à chaud sur aciers). *Bul. Technique du Bureau Veritas (Paris)*, vol. 11, no. 6, June 1929, pp. 122-128, 7 figs.

Detailed report on tests of tensile strength of steel at 450 deg. Cent. for new steam power machinery and piping for new extension of Issy-les-Moulineaux plant of Electric company of Paris; nickel-steel, nickel-chromium, and nickel-chrome-molybdenum steels were tested as well as common steels. (To be continued.)

**STEEL SEGREGATION**

Crystallization and Segregation Phenomena in 1.10 per cent Carbon Steel Ingots of Smaller Sizes. A. Hultgren. *Iron and Steel Inst.—Advance Paper (Lond.)*, no. 6, Sept. 1929, 45 pp., 6 figs.

Results of investigation commenced in 1926 at Soederfors Steel Works in Sweden; deals mainly with 1.10 per cent carbon steel with following average contents: silicon 0.20 manganese 0.30, phosphorus 0.012, and sulphur 0.009 per cent; steel was made in 3.5-ton Electrometall basic electric-arc furnace; molds were generally cold or slightly warm; preheated refractory-lined sink head was used.

**STEEL METALLOGRAPHY**

The Constitution of Steel and Cast Iron Section, F. T. Sisco. *Am. Soc. for Steel Treating—Trans.*, vol. 16, no. 3, Sept. 1929, pp. 435-451, 19 figs.

**STEEL STRESSES**

Permanence of Dimensions Under Stress, W. H. Hatfield. *Metallurgist (Supp. to Engineer, Lond.)*, July 26, 1929, pp. 107-108.

Initial movement occurring in loading, prior to settling down of test piece, is obviously important item in practice, although degree of importance will naturally depend upon exact conditions of use; time effect; internal stresses, left in piece after heat treatment or machining, are obvious cause for obtaining permanent set with relatively low stresses; true behavior of material. (To be continued.)

**TESTING MACHINES**

**FATIGUE.** A New High-Speed Fatigue Testing Machine, H. S. Rowell. *Engineering (Lond.)*, vol. 128, no. 3315, July 26, 1929, pp. 97-98, 6 figs.

Features of machine designed for endurance testing of such plates as are used in

laminated springs, also for general endurance bend test.

**THERMOCOUPLES**

Resistance of Chromium-Nickel Thermocouples to Molten Metals (Die Widerstandsfähigkeit von Chromnickel-Thermoelementen gegenüber Metallschmelzen), R. Hase. *Zeit. fuer Metallkunde (Berlin)*, vol. 21, no. 6, June 1929, pp. 200-203, 10 figs.

Results of tests to determine behavior of chromium-nickel thermocouples in molten bronze and brass; use of protective tubes.

**TUNGSTEN CARBIDE ALLOY**

**MICROSTRUCTURE.** Preparation of Microsections of Tungsten Carbide, S. L. Hoyt. *Am. Soc. Steel Treating—Preprint*, for mtg. Sept. 9-13, 1929, no. 20, 6 pp., 3 figs.

Discussion of method of preparing cemented tungsten carbide for metallographic examination; procedure described has proven to be rapid and efficacious; steps are: smooth grind with silicon carbide wheel grade 60-1 or 100T; lap on corner disk with 140 to 200 mesh boron carbide moistened with kerosene; lap surface with No. 7 diamond dust; time of lapping has been found quite important.

**TUNGSTEN STEEL**

Investigations of Tungsten Steel (Untersuchungen ueber Wolframstahl), W. Zieler. *Archiv fuer das Eisenhuettenwesen (Duesseldorf)*, vol. 3, no. 1, July 1929, pp. 61-78, 66 figs.

Review of most important research and literature; structure of tungsten steel; confirmation of solubility curve of Oberhoffer, Daeges, and Rapatz, as well as of stable and metastable system iron-tungsten-carbon of Hultgren; dilatometric behavior of tungsten steels; hardening capacity; influence of annealing; improvement of tungsten steel by chromium addition.

**WIRE**

**MAGNETIC ANALYSIS.** The Magnetic Analysis of Wire, J. S. Hoffhine. *Wire*, vol. 4, no. 7, July 1929, pp. 228-229 and 245, 2 figs.

New simplified method makes possible rapid non-destructive tests of defects on production basis.

**X-RAY ANALYSIS.** X-Ray Investigation of Wire, F. C. Elder. *Wire*, vol. 4, no. 7, July 1929, p. 236, 2 figs.

Experiments with studies of crystallization in low carbon specimens promise possibilities of wider application for test purposes. Paper presented before Am. Iron and Steel Inst.

**WIRE, STEEL**

**SPECIFICATIONS.** British Standard Specification for Hard Drawn Steel Wire for Concrete Reinforcement. *Brit. Eng. Standards Assn.—Specification (Lond.)*, no. 165, July 1929, 14 pp., 6 figs.

Specification covering definitions, process of manufacture, quality of finished wire, selection and number of tensile and cold-bend test pieces, tests by chemical analysis and at works, purchaser's tests elsewhere, inspection, margin over and under specified dimensions, calculation of weight, identification



of wire, manufacturer's certificate, non-compliance with tests and requirements, delivery, rejection after delivery and arbitration.

#### WROUGHT IRON

**MANUFACTURE.** Byers New Process for Wrought Iron, J. Aston. *Iron Age*, vol. 124, no. 6, Aug. 8, 1929, pp. 341-344, 4 figs.

Description of new process in operation of plant of A. M. Byers Co., Warren, Ohio, whereby 200 tons of wrought iron are produced daily; puddler eliminated by steel plant equipment; mechanical admixture of base metal and slag, one solid and other liquid.

New Methods Promote Use of Wrought Iron, E. C. Kreutzberg. *Iron Trade Rev.*, vol. 85, no. 3, July 18, 1929, pp. 133-135 and 146, 7 figs.

Description of methods employed by Wrought Iron Company of America for manufacturing wrought iron; rotary puddling furnaces make materials more available for their own field; methods of building markets.

#### WROUGHT IRON ANALYSIS

Determination of Slag and Oxides in Wrought Iron. *Chem. Age (Lond.)*, vol. 21, no. 523, July 6, 1929, p. 3.

Slag usually occurs in streaks elongated in direction of working; slag inclusions and form they take endow material with very characteristic mechanical properties; in course of investigation, Wescott, Eckert and Einert examined many of existing methods and developed modification. See reference to their paper, published in *Indus. and Eng. Chem.*, Nov. 1927, in *Eng. Index* 1927, p. 894.

#### WROUGHT IRON PROPERTIES

Comparative Properties of Wrought Iron Made by Hand-Puddling and by "Aston" Process, H. S. Rawdon and O. A. Knight. *Metals and Alloys*, vol. 1, no. 2, Aug. 1929, pp. 46-55 and (discussion), pp. 55-56, 13 figs.

Need for more economical methods in manufacture of wrought iron is pointed out; explanation is given of Aston process; results of chemical analysis of metal at different stages of process are given; results obtained in comparison with properties of iron made by hand puddling and by Aston process.

#### WROUGHT IRON SPECIFICATIONS

Wrought Iron Bars. *Am. Ry. Assn.—Signal Section Proc.*, vol. 27, no. 1, Aug. 1929, pp. 241-243.

Recommended specifications for wrought iron bars for railway signaling.

#### X-RAY ANALYSIS

X-Ray Metallography in 1929. G. I. Clark. *Metals and Alloys*, vol. 1, no. 3, Sept. 1929, pp. 98-111, 13 figs.

Results of analysis of ultimate crystalline constitutions of metals; allotropic modifications; general relationships of structure and properties; analysis of alloy systems; mechanism of solid solution; distinction between "Chemical Compound" and "Solid Solu-

tion"; alloy steels; crystal structure and magnetism. Bibliography. (Continuation of serial.)

Evaluation of Powder X-Ray Photographs (Beiträge zur Auswertung von Debye-Scherrer-Aufnahmen), G. Kettmann. *Zeit. fuer Physik (Berlin)*, vol. 53, no. 3-4, Feb. 12, 1929, pp. 198-209, 9 figs.

Describes graphical method of evaluating powder photographs when cell size is not known; it depends upon extrapolation of curve and its special features are independence of knowledge of fundamental lattice constants and of systematic errors.

Glossary of Terms Used in X-Ray Metallography, L. W. McKeehan. *Am. Soc. for Steel Treating—TRANS.*, vol. 16, no. 3, Sept. 1929, pp. 461-473.

Glossary is primarily intended for those who, while not themselves conducting experiments or tests in which X-rays are used, have occasion to consult publications in this field; it is revised and extended and is supplemented by list of German terms with their English equivalents.

Allotropy and the Determination of Densities by Means of X-Rays. N. H. Kolkmeyer. *Faraday Soc.—Trans.*, vol. 25, no. 98, July 1929, pp. 392-397, 4 figs.

Results obtained from using spectrographic method for crystals of few mm. and also by powder method; special features of powder method apparatus is pointed out.

The Determination of Particle Size By the Use of X-Rays, H. Mark. *Faraday Soc.—Trans.*, vol. 25, no. 98, July 1929, pp. 387-389.

Formulas are developed using Bragg's law as basis.

An Integrating Microphotometer for X-Ray Crystal Analysis. W. T. Atsbury. *Faraday Soc.—Trans.*, vol. 25, no. 98, July 1929, pp. 397-401, 2 figs.

Two main lines of examination required are discussed; identification, by means of direction, of beams diffracted by crystal, and determination of intensities of diffracted beams with respect to that of beam incident on crystal.

Author takes up heat treatment of carburized steel and its effect on structure; effect of cooling slowly from carburizing temperature, quenching directly from box, and regenerative quenching to harden case and refine core; surface hardening of steel is briefly described; brief view of normal and abnormal carburizing steels.

**AUSTENITE DECOMPOSITION.** Austenite and Its Decomposition, A. Sauveur. *Am. Soc. Steel Treating—Preprint for mtg.*, Sept. 9-13, 1929, no. 17, 20 pp., 12 figs.

Mechanism of transformation of austenite remains same whether cooling be slow, resulting in pearlitic structure, or rapid, producing so-called martensitic structure; transformation brought about by alpha phase locating itself at grain boundaries and along some of crystallographic planes of mother austenite; in author's opinion this alpha constituent answers conceptions of troostite; significant differences between hardening of steel after rapid cooling and hardening of some nonferrous alloys after quenching followed by aging.



December

ecture and  
continuation

Photographs  
on Debye-  
mann. *Zeit.*  
p. 3-4, Feb.

evaluating  
size is not  
polation of  
e independ-  
ental lattice  
rs.

-Ray Metal-  
m. *Soc. for*  
no. 3, Sept.

d for those  
ducting ex-  
X-rays are  
publications  
extended and  
erman terms

tion of Den-  
N. H. Kolk-  
vol. 25, no.  
gs.

ing spectro-  
few mm. and  
l features of  
inted out.

e Size By the  
uraday *Soc.*—  
1929, pp. 387.

y Bragg's law

meter for X-  
T. Atsbury.  
no. 98, July

ation required  
by means of  
y crystal, and  
of diffracted  
beam incident

tment of car-  
on structure;  
m carburizing  
tly from box.  
to harden case  
ening of steel  
view of normal  
els.

COMPOSITION.  
mposition, A.  
*Treating—Pre-*  
1929, no. 17,

ion of austenite  
oling be slow,  
ture, or rapid,  
sitic structure;  
by alpha phase  
larities and along  
anes of mother  
tion this alpha  
ons of troostite:  
en hardening of  
d hardening of  
fter quenching

of wire, manufacturer's certificate, non-compliance with tests and requirements, delivery, rejection after delivery and arbitration.

#### WROUGHT IRON

**MANUFACTURE.** Byers. New Process for Wrought Iron, J. Aston. *Iron Age*, vol. 124, no. 6, Aug. 8, 1929, pp. 341-344, 4 figs.

Description of new process in operation of plant of A. M. Byers Co., Warren, Ohio, whereby 200 tons of wrought iron are produced daily; puddler eliminated by steel plant equipment; mechanical admixture of base metal and slag, one solid and other liquid.

New Methods Promote Use of Wrought Iron, E. C. Kreutzberg. *Iron Trade Rev.*, vol. 85, no. 3, July 18, 1929, pp. 133-135 and 146, 7 figs.

Description of methods employed by Wrought Iron Company of America for manufacturing wrought iron; rotary puddling furnaces make materials more available for their own field; methods of building markets.

#### WROUGHT IRON ANALYSIS

Determination of Slag and Oxides in Wrought Iron. *Chem. Age (Lond.)*, vol. 21, no. 523, July 6, 1929, p. 3.

Slag usually occurs in streaks elongated in direction of working; slag inclusions and form they take endow material with very characteristic mechanical properties; in course of investigation, Wescott, Eckert and Einert examined many of existing methods and developed modification. See reference to their paper, published in *Indus. and Eng. Chem.*, Nov. 1927, in *Eng. Index* 1927, p. 894.

#### WROUGHT IRON PROPERTIES

Comparative Properties of Wrought Iron Made by Hand-Puddling and by "Aston" Process, H. S. Rawdon and O. A. Knight. *Metals and Alloys*, vol. 1, no. 2, Aug. 1929, pp. 46-55 and (discussion), pp. 55-56, 13 figs.

Need for more economical methods in manufacture of wrought iron is pointed out; explanation is given of Aston process; results of chemical analysis of metal at different stages of process are given; results obtained in comparison with properties of iron made by hand puddling and by Aston process.

#### WROUGHT IRON SPECIFICATIONS

Wrought Iron Bars. *Am. Ry. Assn.—Signal Section Proc.*, vol. 27, no. 1, Aug. 1929, pp. 241-243.

Recommended specifications for wrought iron bars for railway signaling.

#### X-RAY ANALYSIS

X-Ray Spectrography in Steel, M. E. Clark. *Metals and Alloys*, vol. 1, no. 3, Sept. 1929, pp. 98-111, 13 figs.

tion"; alloy steels; crystal structure and magnetism. Bibliography. (Continuation of serial.)

Evaluation of Powder X-Ray Photographs (Beitraege zur Auswertung von Debye-Scherrer-Aufnahmen), G. Kettmann. *Zeit. fuer Physik (Berlin)*, vol. 53, no. 3-4, Feb. 12, 1929, pp. 198-209, 9 figs.

Describes graphical method of evaluating powder photographs when cell size is not known; it depends upon extrapolation of curve and its special features are independence of knowledge of fundamental lattice constants and of systematic errors.

Glossary of Terms Used in X-Ray Metallography, L. W. McKeehan. *Am. Soc. for Steel Treating—Trans.*, vol. 16, no. 3, Sept. 1929, pp. 461-473.

Glossary is primarily intended for those who, while not themselves conducting experiments or tests in which X-rays are used, have occasion to consult publications in this field; it is revised and extended and is supplemented by list of German terms with their English equivalents.

Allotropy and the Determination of Densities by Means of X-Rays. N. H. Kolkmeijer. *Faraday Soc.—Trans.*, vol. 25, no. 98, July 1929, pp. 392-397, 4 figs.

Results obtained from using spectrographic method for crystals of few mm. and also by powder method; special features of powder method apparatus is pointed out.

The Determination of Particle Size By the Use of X-Rays, H. Mark. *Faraday Soc.—Trans.*, vol. 25, no. 98, July 1929, pp. 387-389.

Formulas are developed using Bragg's law as basis.

An Integrating Microphotometer for X-Ray Crystal Analysis. W. T. Atsbury. *Faraday Soc.—Trans.*, vol. 25, no. 98, July 1929, pp. 397-401, 2 figs.

Two main lines of examination required are discussed; identification, by means of direction, of beams diffracted by crystal, and determination of intensities of diffracted beams with respect to that of beam incident on crystal.

Author takes up heat treatment of carburized steel and its effect on structure; effect of cooling slowly from carburizing temperature, quenching directly from box, and regenerative quenching to harden case and refine core; surface hardening of steel is briefly described; brief view of normal and abnormal carburizing steels.

**AUSTENITE DECOMPOSITION.** Austenite and Its Decomposition, A. Sauveur. *Am. Soc. Steel Treating—Preprint for mtg.*, Sept. 9-13, 1929, no. 17, 20 pp., 12 figs.

Mechanism of transformation of austenite remains same whether cooling be slow, resulting in pearlitic structure, or rapid, producing so-called martensitic structure; austenite is shown locating itself at grain boundaries and along some of crystallographic planes of mother

d  
n  
  
s  
e-  
t.  
o.  
  
g  
ot  
of  
l-  
ce  
  
l-  
or  
ot.  
  
se  
x-  
re  
ms  
nd  
ms  
  
en-  
lk-  
no.  
  
ro-  
nd  
of  
  
the  
-  
87-  
  
aw  
  
X-  
ry.  
uly  
  
red  
of  
and  
ted  
ient  
  
car-  
are;  
ting  
box,  
case  
steel  
mal  
  
ON.  
A.  
Pre-  
17,  
  
ente  
low,  
apid,  
ure;  
-  
ling  
other